TRANSACTIONS
American Society for Steel Treating

VOL. X JULY, 1926 NO. 1

MINUTES OF THE MEETING OF THE BOARD OF DIRECTORS

HOTEL BOND, HARTFORD, MAY 19, 1926—11:15 A. M.

Present:

PRESIDENT BIRD
DR. ZAY JEFFRIES
W. S. BIDLE
F. G. HUGHES
F. E. MCCLEARY
R. G. GUTHRIE
HYMAN BORNSTEIN
J. F. HARPER
W. H. EISENMAN

UPON motion by Mr. Hughes, seconded by Mr. Bornstein and unanimously carried, the minutes of the previous meeting were approved.

It was moved by Mr. Guthrie, seconded by Mr. Harper and unanimously carried that local chapters be relieved of all responsibilites relative to convention activities, and that the National Society so organize as to assume all responsibilities and make all arrangements and plans for conventions, irrespective of the local chapter, which will function only in an advisory capacity.

Upon motion by Dr. Jeffries, seconded by Mr. Hughes and carried, it was moved that subject to approval by the Detroit chapter, the 1927 convention and exposition should be held in Detroit, preferably during the week beginning September 19th.

It was moved by Mr. Bornstein, seconded by Mr. Hughes and unanimously carried that the January, 1927, Sectional Meeting should be held in Washington, D. C.

It was the consensus of opinion of the Board that Milwaukee would be a desirable place for the Spring Sectional Meeting in 1927.

Reports of committees were then received for the 1926 convention.

Upon motion by Mr. Eisenman, seconded by Mr. Hughes and unanimously carried the sum of $4000.00 was appropriated to the Chicago Convention Committee.

F. E. R.

The secretary reported that 76,000 square feet of space had been reserved to date for a total of $80,150.00 and that about 5000 square feet remained unreserved in the Steel Section.

It was the consensus of opinion that the American Society for Steel Treating should have an educational exhibit at the Chicago Convention and the National Office was instructed to prepare such an exhibit.

Upon motion by Mr. Bornstein, seconded by Mr. Hughes and unanimously carried, a committee composed of President Bird, W. S. Bidle, Dr. Jeffries and W. H. Eisenman was elected to secure the services of an assistant secretary and such other help as might be necessary for the organization at the National Office.

President Bird gave the result of his visits with the executive committee of the chapters and their attitude toward an increase of activities in the non-ferrous field. Twenty of the chapters visited were in favor of further activities in this field, four were noncommittal, while one chapter was opposed.

President Bird also reported the result of his conference at Washington, held during the meeting of the Advisory Committee, on Non-Ferrous Subjects of the Bureau of Standards, at which many members of the Institute of Metals Division of the A. I. M. M. E. and the A. S. S. T. were present. At this meeting a splendid spirit of co-operation was shown and assurance of further co-operation by the Institute of Metals in the preparation of data sheets, etc., was given.

It was moved by Dr. Jeffries, seconded by Mr. Hughes and unanimously carried, that the A. S. S. T. increase its activities in connection with non-ferrous metals.

Upon motion by Mr. Hughes, seconded by Mr. Guthrie and unanimously carried, it was moved that power be delegated to the National Office to organize members into a group in centers where a desire for such organization is recognized and where distance renders attendance at present chapter meetings impracticable, these groups be organized by and with the consent of the Board of Directors and that such groups shall have the same type of organization as the chapter and shall function similar to the chapter except that the group will have no representative on the National Nominating Committee.

Upon motion by Mr. Bornstein, seconded by Mr. Guthrie and

unanimously carried, charters were issued to Fort Wayne and Montreal as groups.

It was moved by Mr. Hughes, seconded by Mr. Guthrie and unanimously carried that student groups (junior members) might be sponsored by local chapters or by authority of the Board of Directors.

The Case Group organized among junior members at Case School of Applied Science and sponsored by the Cleveland chapter was so recognized.

The secretary presented a report on metallurgical educational activities in the various centers in which the A. S. S. T. has chapters as well as educational activities in which the local chapters are participating. This was accepted as a report of progress and President Bird was authorized to make arrangements for a Metallurgical Educational Luncheon, to be held during the annual Convention of the Society at Chicago.

Upon motion by Dr. Jeffries, seconded by Mr. Bornstein and unanimously carried, J. H. Gibboney's appointment by President Bird as a member of the Recommended Practice Committee was confirmed.

President Bird reported the resignation of Robert Schenck as a member of the Recommended Practice Committee and the secretary was instructed to send Mr. Schenck a letter of regret and also an expression of thanks for his activity while serving on that committee.

Upon motion by Dr. Jeffries, seconded by Mr. Bornstein and unanimously carried, the appointment by President Bird of H. M. Northrup, metallurgist, Hudson Motor Car Company, a member of the Recommended Practice Committee, was confirmed.

Mr. Harper presented a progress report of the Recommended Practice Committee and particularly of the committee on Heat Treatment Definitions. This report was accepted.

The question of free distribution of the Society's publication to libraries and engineering societies did not meet with full approval of the Board and it was the consensus that no gratis distribution of the Society's publications should be made but that where possible and desirable, an exchange of publications should be effected.

The question of a Membership Directory was given consideration and in view of the cost (approximately $3000.00), it was

deemed inadvisable to prepare a Membership Directory at this time.

Consideration was given to the question as to whether or not it would be desirable to change the Constitution so as to permit associate members to be eligible as vice-chairmen or chairmen of local chapters.

Upon motion by Mr. Hughes, seconded by Mr. McCleary and unanimously carried, it was moved that nothing be done to change the class or privileges of associate members.

It was the consensus of opinion of the Board that it would be desirable that a letter of invitation, together with an executive pass for the Convention and Exposition to be held in Chicago, should be sent to those executives whose names have been forwarded by local chapters and who had already received the executive pamphlet.

It was moved by Mr. Eisenman, seconded by Mr. Hughes and unanimously carried, that the National Meetings and Papers Committee should prepare the technical programs for all sectional meetings after consultation with the local chapter in the city in which the sectional meeting is to be held.

The question of free data sheets to new members, bound TRANSACTIONS free to new members for the first year and payment of expenses of delegates to National Conventions were not looked upon favorably by the Directors.

Upon motion by Dr. Jeffries, seconded by Mr. Guthrie and unanimously carried, authority was granted President Bird to appoint such representatives on co-operative committees as may occur and require action.

It was decided that the next meeting of the Board of Directors should be held in Chicago during the annual convention.

Upon motion properly made, seconded and carried, the meeting adjourned at 5:45 P. M.

EXTENSION MEETINGS

THE past year has shown a marked increase of activity on the part of the chapters of the A. S. S. T. in the holding of extension meetings. While this has been engaged in for a number of years past, the plan has received renewed impetus since the visit of President Bird with chapter executives, at which time he pointed out the responsibility resting upon local chapters for in

creasing their services to nearby communities and also for the purpose of bringing to the attention of those who should be interested, the advantages of membership in the Society.

Without attempting to enumerate all of the extension meetings that have been held, it is interesting to note a few of the following: Boston at Lynn; Buffalo at Lockport; Cincinnati at Hamilton and Newport; Detroit at Ann Arbor; Golden Gate at Pittsburg; Hartford at New Britain; Lehigh Valley at Reading; Los Angeles at Torrence; New Haven at Waterbury and Bridgeport; New York at Newark; Rhode Island at New Bedford; Schenectady at Albany; Toronto at Hamilton and Washington at Baltimore.

For the coming year the Detroit chapter also plans to hold four extension meetings during the year at Ann Arbor, where the students of the University of Michigan, as well as the members of the Society in that vicinity, will have the opportunity to obtain benefit, and realize the excellent work being done by the local chapter.

CHICAGO AND THE CONVENTION

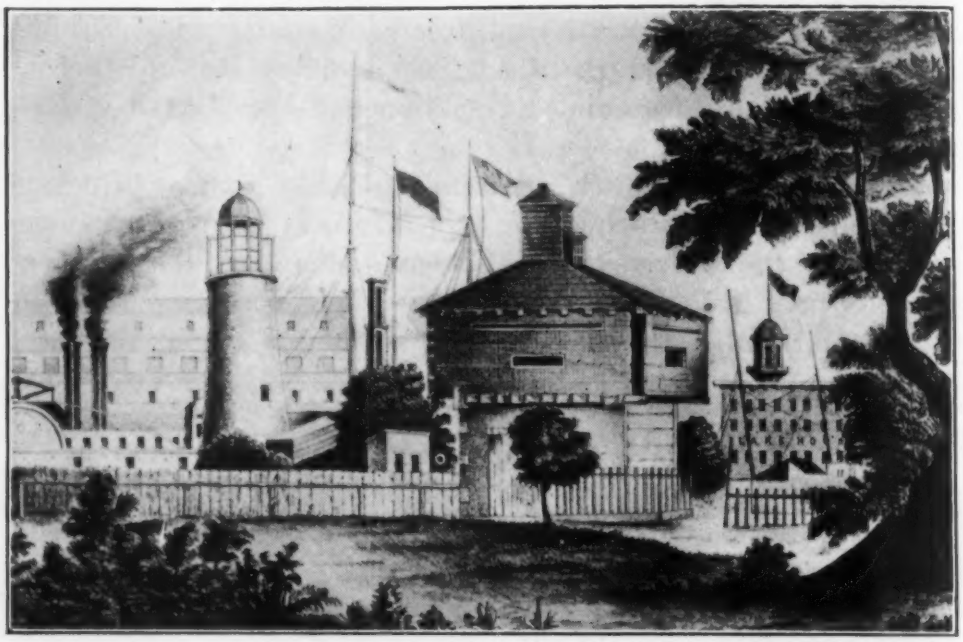
CHICAGO, where the American Society for Steel Treating Convention and Exposition will be held the week of September 20, is called the magic city of the world. Its rapid growth is phenomenal and probably unparalleled. The city proper is less than one hundred years old, yet it is now third largest in the world, with an area of about 200 square miles and a population of approximately 3,000,000. Its growth is estimated at about 60,000 per year.

It is situated 190 miles northwest of the center of population and is accessible from every direction,—on three sides by innumerable automobile roadways and the 38 railroads which terminate in Chicago; and on its easterly side by more than 22 miles of level lake front, which is adequately equipped with dock and railroad facilities.

La Salle as early as 1682 prophesied that it would be the gateway of a nation and rank foremost in commerce. He, doubtless, based his judgment on its strategic position and geographical advantages. The development of the Lakes to the Gulf and the Great Lakes-St. Lawrence Waterways will result in Chicago becoming a highly important inland seaport.

Situated in the north temperate zone, its climate is delightful and of such a bracing nature that its citizens are energetic, ambitious and persevering. Its business principles are positive and sound.

But in imagination let us go back to the early days. The boy who in answer to the teacher's inquiry facetiously replied that Chicago was at the bottom of Lake Michigan, deserved much credit for his answer, for geologists are of the opinion that that was pre-



Fort Dearborn in Its Early Days

cisely the position it occupied in the long, long ago. The waters of Lake Michigan entirely covered this portion of the earth in the dim distant past. As the lake gradually receded, it left behind it a small river with two branches. Vegetation upon the marshy banks of the river grew profusely and here the Indians found the wild garlick, leek or onion growing in great abundance. They called this creek or river Checaugua. And here residing along this river the early French explorers found about 20,000 Indians. These were the Illini and their most famous chief was Checaugua, meaning the strong leader. The word Checaugua seems to have many interpretations but they always carry the underlying idea of strength. Though of lowly origin and pungent odor, nevertheless the little wild leek bulbs were hardy and nourishing and sustained many an early pioneer who might have died of hunger but for

them. Yesterday Chicago furnished the lowly leek to her guests, today she is a great metropolis feeding a hungry world.

In 1675, Father Marquette found a lively trade in fur in progress between the Indians and the French-Canadians in the vicinity of what is now Chicago. This was on the route from Canada to Louisiana.

In 1763 the territory of what is now Illinois was ceded by France to Great Britain and in 1778 Captain Clark of Virginia took it by force from Great Britain.

While the Indians and early pioneers called the river Checaugua, the earliest settlement was known as Fort Dearborn, named after the then secretary of war. The first fort was built in 1803 on the south side of the river where it enters the lake. It was constructed by a troop of United States soldiers under Captain John Whistler, whose grandson was a distinguished painter of our day. The fort consisted of two blockhouses which were enclosed by a palisade of wooden pickets, fortified with three pieces of light artillery. A subterranean passage extended from the parade ground to the river. This furnished a place of hiding and a way of escape in an emergency for the occupants of the four cabins which were erected the first year and in time of siege it could be used to supply the garrison with water.

In 1812, when war was declared by the United States against Great Britain, Fort Dearborn was evacuated, the inhabitants going to Fort Wayne at Detroit. A second Fort Dearborn was built in 1816. In 1833 the town of Chicago was incorporated and in 1837 it became a city and had a population of about 4000. It became a port of entry in 1846. In 1831 the first post office was opened and mail brought from Detroit twice a week on horseback. Today, less than a century later, an average of 5,000,000 pieces of mail is handled daily and its annual post office receipts amount to $55,000,000.

Though Nature was lavish with her gifts to Chicago and prepared a place for a wonderful city by the lake, yet another and extremely vital factor enters into her greatness. It is said that the civilization of a people is an expression of that people and is the index of the character of that people. Let us remember that Chicago was originally set on a swamp below the level of Lake Michigan and that it was badly crippled and handicapped by the great fire in 1871 and that prior to that time it had three fires

which destroyed much of its wealth, yet human ingenuity, enterprise and energy conquered. Their motto, "Chicago first, last and all the time," is good constructive psychology and surely they have justified it by their deeds. Thus this great city is an embodiment of its people. Now the question arises, who are its people, for nearly everything of note which has happened in Chicago is within the memory of those living, so young is Chicago. Her citizens are a combination of the best blood of New England and the South. Pioneers are ambitious people, daring, courageous, highly imaginative, hopeful, strong, persevering. From New England and Virginia came these marvelous men and women who settled here. In the State of Illinois these two great peoples met—and the result is Chicago. The unswervable bravery of the Pilgrim fathers and the gallant and chivalrous spirit of the southern cavaliers produced a new people, and a new city—Chicago.

Idealism is deep-rooted in its people. They dream dreams and *make* them come true. The abounding faith of the citizens in their own city is one of the greatest assets of Chicago.

Situated centrally as it is, this city has for many years been a favorite convention city. Here may be seen daily, people from all over the world in convention. It is the great meeting place of people and ideas.

It is here in this great city of Chicago that members of the American Society for Steel Treating will hold their eighth annual convention and exposition September 20-24.

This enormous exposition, the largest of its kind ever held, will be located on the Municipal Pier, which is situated at the foot of Grand Avenue. More than 75,000 square feet of floor space, exclusive of aisles, will be devoted to the exhibition of the many commodities, machinery and equipment which enter into the production of steel and its heat treatment.

The amount of floor space to be occupied at this exposition will be nearly double that which was occupied at Cleveland last year.

An excellent technical program has been arranged by the Meetings and Papers Committee which, together with the many other activities that go to make up the annual convention, promise well that the Chicago Convention in 1926 will excel all previous conventions.

The Drake Hotel will be the headquarters of the convention.

MACROSCOPIC EXAMINATION OF IRON AND STEEL

By F. P. Gilligan and J. J. Curran

Abstract

This paper describes the deep-etch method of testing or inspecting steel bars, billets or finished parts. It discusses in detail the procedure to be followed in using this method of test and the merits of the various acid mixtures used in detecting the different types of defects occurring in steels which may be revealed by this type of test. A comparison between this test and the microscopic test is made.

The authors present a large number of photographs of specimens tested by this method and describe the nature of the defects as revealed.

TWO different classes of structures are commonly dealt with in the examination of iron and steel—microstructure and macrostructure. The microstructure is made visible only by the aid of the microscope, hence its name. The examples of microstructure which appear in print, usually at a magnification of 100 diameters, represent small areas of the metal, rarely more than 0.05 inches in diameter, magnified to show the details of the internal structure. If the magnification is carried much higher, to 2500 diameters, for instance, the actual area shown in the photomicrograph will probably be not over 0.002 inches in diameter.

As opposed to this, the macrostructure is that structure which can be made visible to the unaided eye. In macroscopic examination, the area observed may be of any size within the usual range of vision. It lacks the extreme detail shown by microscopic examination, but it has the advantage of showing the aggregate effect of these details, and on a scale which shows their relationship to the form and size of the section of material being examined.

Microscopic examination requires careful preparation of specimens, and the operations involved are such that extremely large sections cannot be prepared conveniently for examination. Fur-

A paper presented before the Buffalo Sectional Meeting, January, 1926, by J. J. Curran. Of the authors F. P. Gilligan is secretary and J. J. Curran is metallurgist with the Henry Souther Engineering Company, Hartford, Conn. Both authors are members of the Society, the former being a past president.

ther, considerable experience and skill is necessary for accurate interpretation of the structures observed.

Macroscopic examination, on the other hand, does not usually require such careful preparation, and complete cross-sections of the average steel object can be easily prepared for examination. The macrostructure, when properly developed, is usually practically self-explanatory, only a general idea of the nature of the metal and its method of manufacture being necessary to permit reasonably accurate interpretation.

Macroscopic examination, while it does not supplant the use of the microscope, offers a simple, convenient method for the inspection and study of steel. The object of this paper will be to discuss some of the methods, and to show some of the conditions which may be disclosed by such examination. In order to better understand the conditions which determine the macrostructure, a brief description of the mechanism of solidification of steel and its subsequent working is included.

By reason of the progressive nature of solidification, cast metals, unless extremely pure, are never entirely homogeneous. The presence of small quantities of a second metal or element as an impurity is in most cases sufficient to produce variations in composition from point to point in individual grains or crystals, and from point to point within the casting itself. Subsequent heat treatment of the cast metal will tend to eliminate some of the microscopic variations, and this action will be more complete if the material is mechanically worked at high temperatures, the easily diffused elements of impurities tending to migrate from the concentrated areas to the deficient areas until uniformity is established. Those elements which do not diffuse readily, or which do not enter into solid solution at all, retain their respective positions and concentrations except as modified by the distortion produced in the mechanical working.

Steel, in its simplest form, is essentially an alloy of iron and carbon, containing varying amounts of manganese, silicon, aluminum, phosphorus and sulphur, and alloying metals like nickel, chromium, etc., in the case of alloy steels. Small amounts of oxygen, hydrogen, carbon monoxide, nitrogen, and other gaseous elements are also usually present. Some elements, as manganese and silicon, are essential for the production of a satisfactory steel; while others, such as phosphorus and sulphur, are undesir-

able impurities except in a few instances. Each of these elements, desirable and undesirable, tends to alter the melting and solidification characteristics of the base metal, usually causing the metal to melt or freeze progressively over a range of temperature instead of at a single definite temperature, as in the case of pure iron. As their number increases this effect becomes more pronounced, so that in the average steel, containing from five to ten elements in addition to iron, their effect on the melting and solidification characteristics is profound.

Steel, in the molten condition, ready for pouring, may be considered as a solution in pure iron of any or all of these constituent elements, and containing in suspension solid particles and fluid droplets of silicates, sulphides and oxides. When poured into the mold, solidification starts immediately. The first solid grains or crystals are formed next to the mold surface, and, as heat is extracted from the metal and mold, these crystals grow towards the center. The freezing at first is rapid, then progresses more slowly, due to the slowing up of heat conduction from the metal to the mold and to the outside atmosphere, and due also to the fact that as the content of impurities in the portions still liquid increases, the freezing point is lowered progressively.

At the same time, a contraction in volume is taking place due to the change from liquid to solid state, and due to constant lowering of temperature of the already solid metal. This contraction tends to lower the level of the liquid metal in the solid shell, giving rise to a shrinkage cavity or pipe, unless means are provided to feed in liquid metal to preserve the original liquid level.

During solidification, and during cooling from the freezing point, changes are continually taking place in the solid metal, the most important of which is a reversal of the effects of freezing. That is, the microscopic, and to some extent, the macroscopic segregation produced by the progressive nature of the freezing operation tends to disappear through diffusion. Those elements which are now in solid solution, and capable of easy diffusion, tend to distribute themselves uniformly through the solid metal. The less easily diffusible elements, as phosphorus, dissolved oxygen, and sometimes nickel, do not equalize in this manner, but remain in relatively high concentrations in the portions of the metal which were last to solidify. In these same areas will be

found also the major portion of the silicates, sulphides and oxides, now called sonims, or solid non-metallic impurities—"dirt"—suspended in the molten steel at the time of pouring. If the steel is of the effervescing type, evolving considerable gas during the solidification processes, there will often be found blow-holes partly filled with the highly impure molten metal which was last to solidify, and which was forced into the blow-holes by the pressure of the contracting solid steel.

Annealing of a casting, for example, an ingot, showing these characteristics, promotes further equalization through diffusion, and even those elements which diffuse with difficulty may be affected. The sonims, or dirt, however, cannot diffuse and are immobile in the solid steel, therefore their distribution remains unchanged. The large grains of the original casting may now appear under the microscope to be fairly well broken up by the annealing treatment, but macroscopically the steel will still show the cast or ingot structure due to the presence of the undiffused constituents and the accompanying sonims. Forging and hot rolling of the ingots into blooms, billets and bars promotes further diffusion, depending on the amount of hot work applied, and the ingot structure will usually be completely broken up by such working. Owing to the fact, however, that forging and rolling operations are attended by a reduction of dimensions in one direction only, the incompletely diffused areas are now drawn out in the form of long filaments, parallel to the direction of rolling, giving the bar a "fibrous" appearance when longitudinal sections are examined macroscopically. This "fiber" is determined by the segregation present, that is, a bar made from an ingot which was relatively uniform and free from serious segregation will be relatively free from fiber, whereas a bar made from an ingot originally badly segregated and dirty will show a pronounced fiber.

The macrostructure of the steel is determined by the phenomena just described, and can be disclosed by numerous methods, most of which have their peculiar advantages and disadvantages. These methods will be enumerated and discussed briefly here.

Probably the oldest and simplest method of examining the macrostructure of steel is the fracture test. Another old method of examining steel macroscopically is the sulphur print, in which the distribution of the sulphide inclusions in the steel is disclosed.

Since the sulphide inclusions are usually found in conjunction with other impurities which tend to segregate, the distribution of the sulphides is a fairly reliable indication of the distribution of the other impurities and therefore of the homogeneity of the steel. Iodine solutions are used to indicate the regularity of distribution of elements in solid solution, but chiefly as to carbon and phosphorus. Other methods employ reagents containing copper salts in solution. These show the distribution of the various elements by the manner in which the copper is deposited from the solution on the surface of the specimen. They are best adapted to show segregation of phosphorus and dissolved oxygen. Any or all of these methods will yield valuable data in the hands of an expert, but they require careful preparation and much skill in their application.

The method which we consider the simplest, easiest to apply, and most productive of information to trained and untrained observer alike, is the hot acid etch. In this method the specimen requires no polishing; a fine saw-cut or turned surface is usually satisfactory. Rough grinding will improve the detail, and is desirable when convenient. The specimen is immersed in a strong acid solution and kept at a nearly boiling temperature (about 200 degrees Fahr.) until the surface is properly etched to reveal the structure. The solution may be any of the strong mineral acids, in varying concentrations, usually from 25 per cent up. Hydrochloric acid and mixtures of sulphuric and hydrochloric acid have been used in our practice with best results, in the following proportions:

A.	Hydrochloric acid	—9 parts by volume
	Sulphuric acid	—3 parts by volume
	Water	—1 part by volume
B.	Hydrochloric acid	—7 parts by volume
	Sulphuric acid	—1 part by volume
	Water	—2 parts by volume
C.	Hydrochloric acid	—1 part by volume
	Water	—1 part by volume

All these solutions give practically identical results, solution A giving slightly better detail, with less tendency to rust subsequently, while taking nearly twice as long to produce the re-

sults, as solution C. Solution B represents a compromise between the two, which has given satisfaction.

After etching, the specimens should be washed thoroughly in running water, preferably hot water, and dried quickly by pressing on a soft cloth, then washed with denatured alcohol and dried, first on cloth, then by means of an air blast or breeze of an electric fan. The purpose of these washings is to prevent rusting. With some practice specimens can be prepared which will retain their characteristic appearance for months without rusting. For longer preservation, a coat of thin lacquer may be applied. Rusting is more pronounced with solution C, therefore most care is necessary in washing after its use. If the specimens are not to be photographed or preserved for reference, the alcohol washing may be eliminated, and the specimens dried in the air blast directly after washing and drying on cloth.

The action of the hot etch is simple. Metal that is sound and homogeneous and relatively free from inclusions or dirt will be only slowly and evenly attacked by the acid. If, however, considerable quantities of inclusions are present, these inclusions will be quickly dissolved out, and the solution of the metal will be much more rapid, due to the increased surface thus exposed. As a result, metal containing large amounts of dirt, evenly distributed, will be deeply pitted throughout. Thus a section of high quality crucible steel will be only lightly etched, whereas under the same treatment common screw stock will be deeply pitted. Segregations will be made visible due to differences in solubility of the segregated metal, and also due to the fact that most of the dirt will be concentrated in the segregated area. Laps, seams, or other discontinuities in the metal will be made visible by the attack of the acid on the inclusions usually found in these defects.

The flow of metal in forging and heading operations may be disclosed by means of the hot acid etch applied preferably to longitudinal sections, the changes in direction of the "fiber" being usually quite prominent. The etch may also be applied to show effects produced in the steel after treatment and in service, as for instance, hardening cracks, grinding cracks, fatigue cracks, etc.

In order to show the effects of the hot acid etch, a number of representative plates are included. These illustrate some of the more important uses and features revealed by etching samples

encountered in daily work in the laboratory.

Fig. 1 shows the heterogeneous macrostructure of a bar of wrought iron, as developed by the hot acid etch. The lightly etched portions, at E and F, have the characteristics of steel, whereas the deeply etched and pitted portions are characteristic for true wrought iron.

Fig. 1—Heterogeneous Macrostructure of a Bar of Wrought Iron, as Developed by the Hot Acid Etch. The Lightly Etched Portions, at E and F, have the Characteristics of Steel, whereas the Deeply Etched and Pitted Portions are Characteristic for True Wrought Iron.

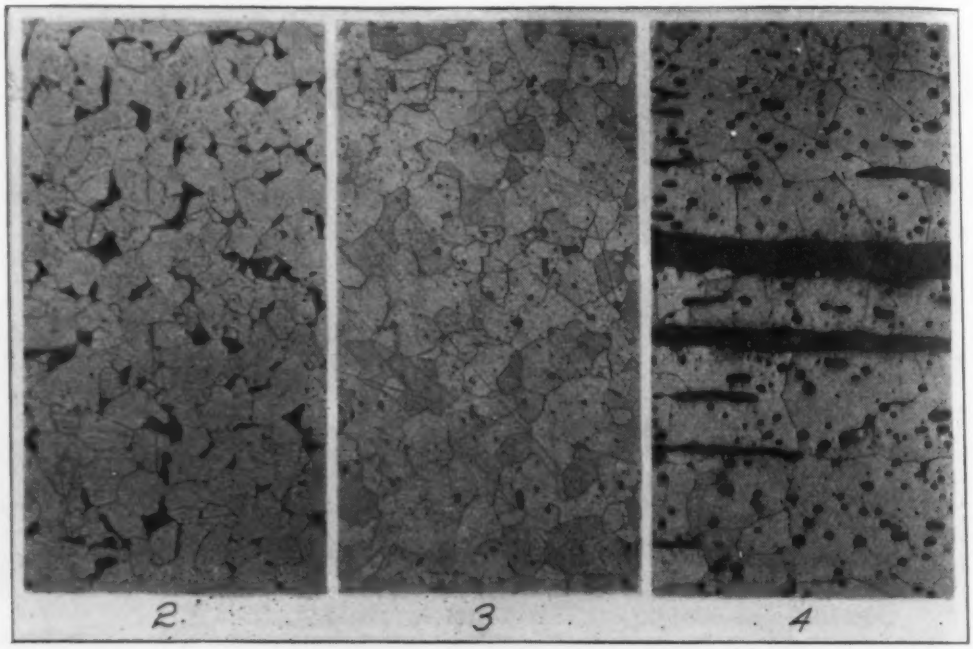
Figs. 2, 3, and 4, show the microstructure of the metal at the points marked E, F, and G, in Fig. 1, and corroborate the indications of the acid etch, i. e., the light etching portions, E and F, have a slagless steely structure, whereas at G the normal wrought iron structure is observed. Thus the hot acid etch may be useful to disclose adulteration of the wrought iron by the addition of steel in the piling operation. The effect of inclusions on the resistance of the metal to the etch is well illustrated by this specimen.

Fig. 5 shows the dendritic or "ingot" structure of a steel casting. The specimen was actually cut from the riser of a casting, and shows the characteristic shrink or pipe.

Fig. 6 shows the same dendritic structure persisting in a

"forged" die blank. The severe reduction in hot working has not sufficed to break up this structure entirely.

Fig. 7 shows the macrostructure of a badly segregated 2½ inch square bar of 0.40 carbon steel. The hot acid etch has developed a picture frame effect in which the outer portion, or



Figs. 2, 3 and 4—Microstructure of the Metal at the Points marked E, F, and G, in Fig. 1, Corroborate the Indications of the Acid Etch, i. e., the Light Etching Portions, E and F, have a Slagless Steely Structure, whereas at G the Normal Wrought Iron Structure is Observed. Thus the Hot Acid Etch may be Used to Disclose Adulteration of the Wrought Iron by the Addition of Steel in the Piling Operation. The Effect of Inclusions on the Resistance of the Metal to the Etch is well Illustrated by this Specimen.

frame, is only lightly etched and the central portion, within the frame, is deeply pitted. A sulphur print of one-half of the cross-section of the same bar shows a similar frame—Fig. 8.

In Figs. 9 and 10 the microstructures of the frame and central portions of Figs. 7 and 8 are compared. The segregated condition of the center of the bar is indicated by the higher volume of pearlite (dark areas) present and the streaks of ferrite (white areas) as compared with the smaller volume of pearlite and more evenly distributed ferrite in the outer portion.

Analyses of drillings taken from the outer or frame portion of the bar, and from the central portion, indicated differences in composition as follows:

	Frame Portion Per Cent	Central Portion Per Cent
Carbon	0.336	0.427
Manganese	0.49	0.55
Phosphorus	0.030	0.052
Sulphur	0.030	0.049

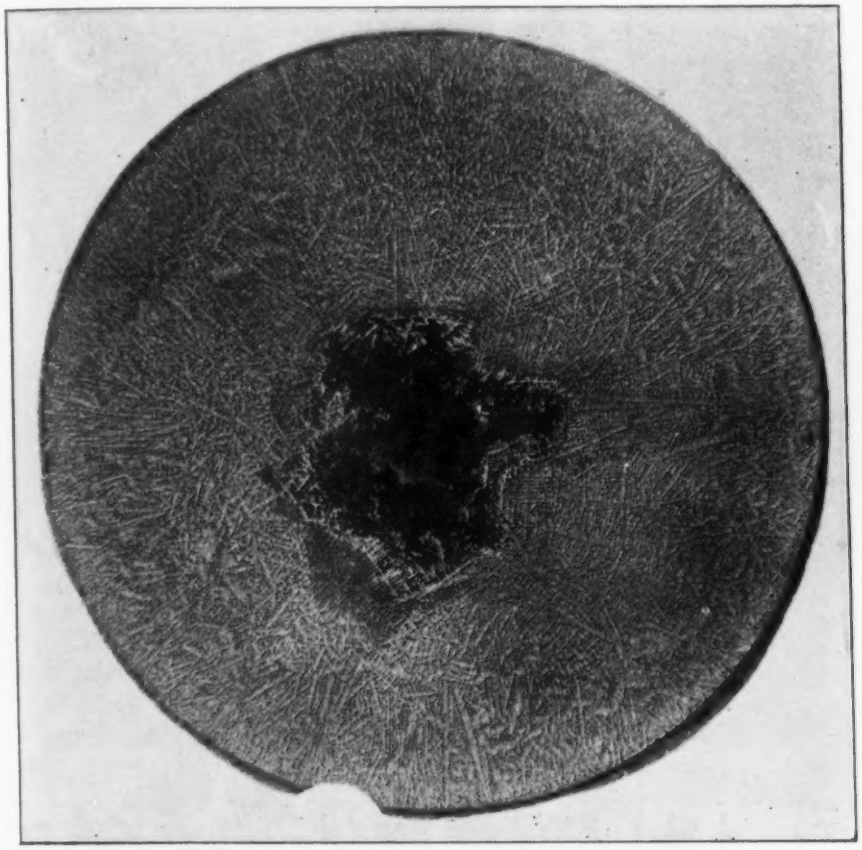


Fig. 5—Dendritic or "Ingot" Structure of a Steel Casting. The Specimen was Actually Cut from the Riser of a Casting, and shows the Characteristic Shrink or Pipe.

Fig. 11 shows an etched section of a case-hardened forging which cracked in hardening. It should be observed that the crack occurred at the point where the unsound segregated center of the bar was brought to the surface in the flash formed in forging. The unsoundness observed in this bar would not ordinarily be considered serious, but for this particular purpose, its effect was much more serious than expected.

Figs. 12 and 13 show the macrostructure of good and bad bars of tool steel, approximately 2⅝ inches in diameter. The

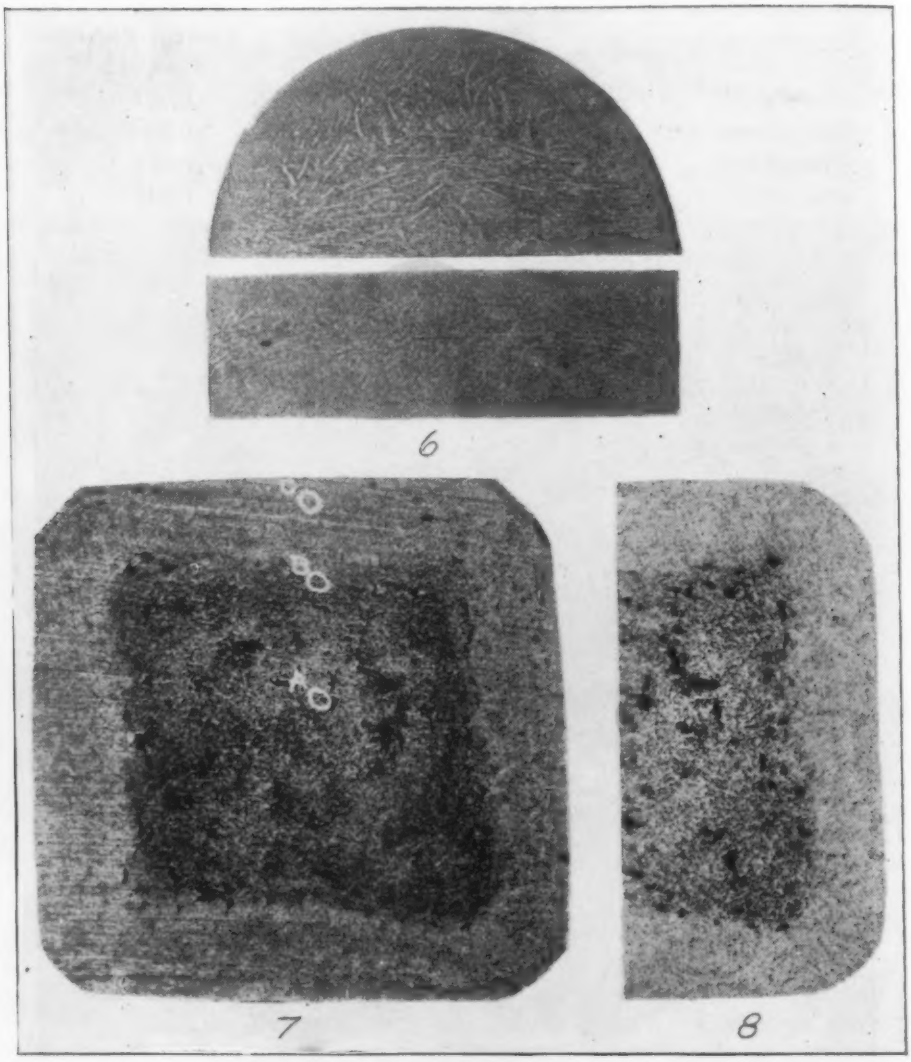
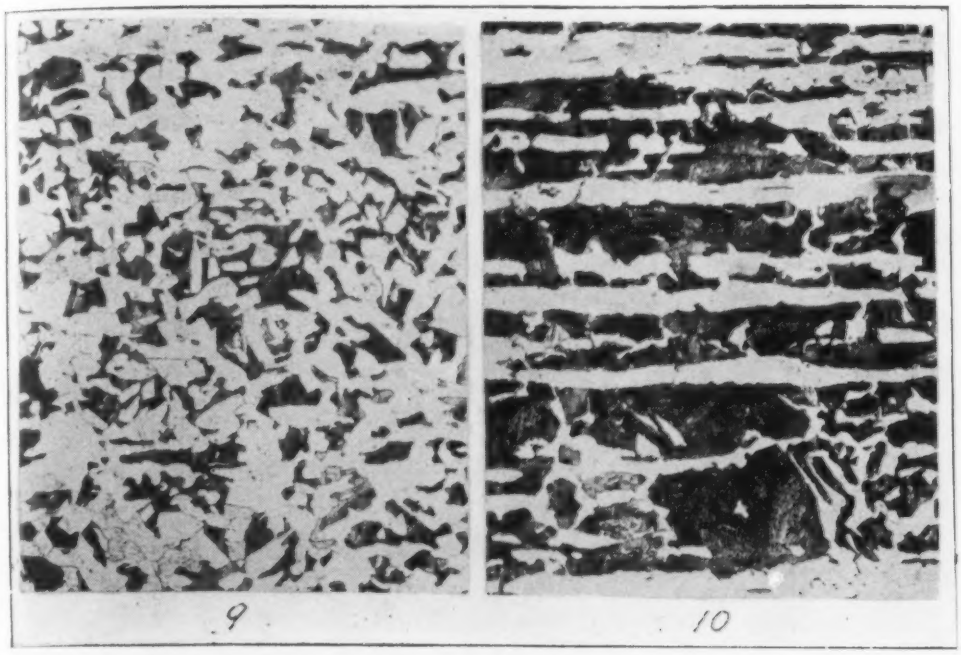


Fig. 6—Dendritic Structure of Fig. 5 Persisting in a "Forged" Die Blank. The Severe Reduction in Hot Working has not Sufficed to break up this Structure Entirely.

Fig. 7—Macrostructure of a Badly Segregated 2½ Inch Square Bar of 0.40 Carbon Steel. The Hot Acid Etch has Developed a Picture Frame Effect in which the Outer Portion, or Frame, is only Lightly Etched and the Central Portion, within the Frame, is Deeply Pitted. A Sulphur Print of One Half of the Cross-Section of the same Bar Shows a Similar Frame—Fig. 8.

good bar shows no appreciable pitting, whereas the bad bar shows many irregular deeply pitted areas. This bad bar, when made into tools, had a record of 100 per cent failure by cracking in hardening.

Fig. 14 shows the decarburized rim on a small bar of high speed tool steel. The depth of decarburization is excessive in this instance. In many cases the hot acid etch will serve to show



Figs. 9 and 10—Microstructures of the Frame and Central Portions of Figs. 7 and 8 are Compared. The Segregated Condition of the Center of the Bar is Indicated by the Higher Volume of Pearlite (Dark Areas) Present and the Streaks of Ferrite (White Areas), as Compared with the Smaller Volume of Pearlite and more evenly Distributed Ferrite in the Outer Portion.

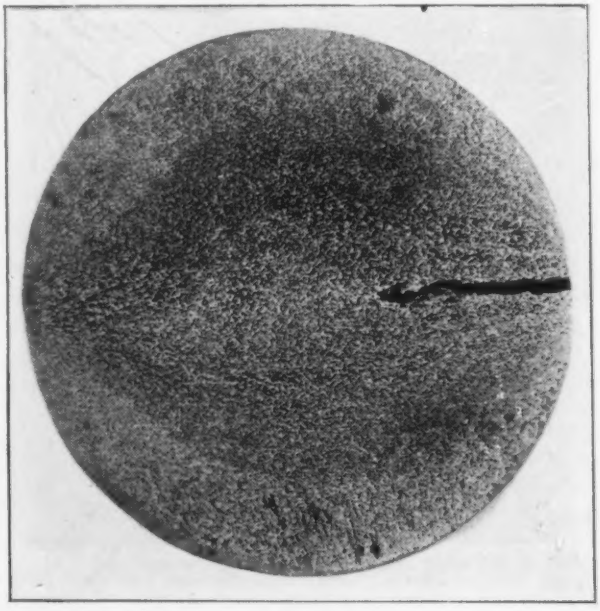
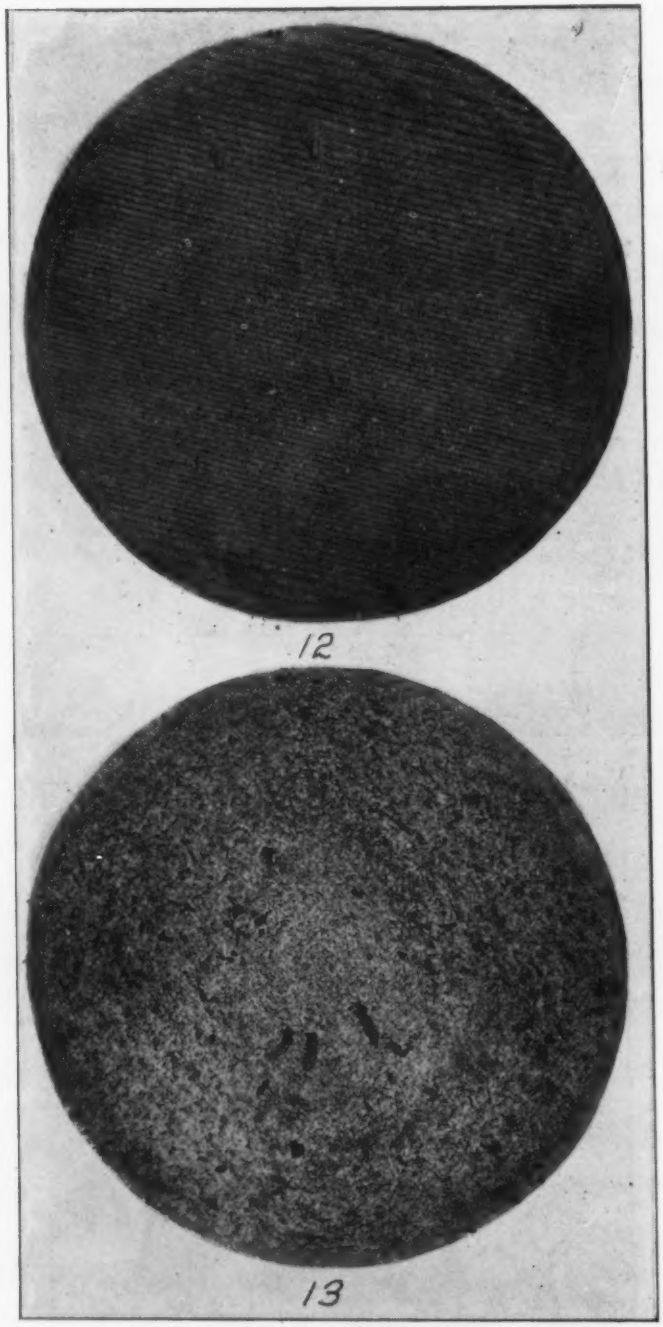


Fig. 11—Etched Section of a Case-Hardened Forging which Cracked in Hardening. It Should be Observed that the Crack Occurred at the Point where the Unsound Segregated Center of the Bar was Brought to the Surface in the Flash Formed in Forging. The Unsoundness Observed in this Bar would not Ordinarily be Considered Serious, but for this Particular Purpose, its Effect was much more Serious than Expected.



Figs. 12 and 13—Macrostructure of Good and Bad Bars of Tool Steel, Approximately 2⅝ Inches in Diameter. The Good Bar Shows no Appreciable Pitting, whereas the Bad Bar Shows Many Irregular Deeply Pitted Areas. This Bad Bar, when Made Into Tools, had a Record of 100 Per Cent Failure by Cracking in Hardening.

the amount which must be removed from the surface of tool steel to reach the normal metal.

Fig. 15 shows characteristic flow or strain lines in a forged ring. These merely show the direction of flow of the metal and

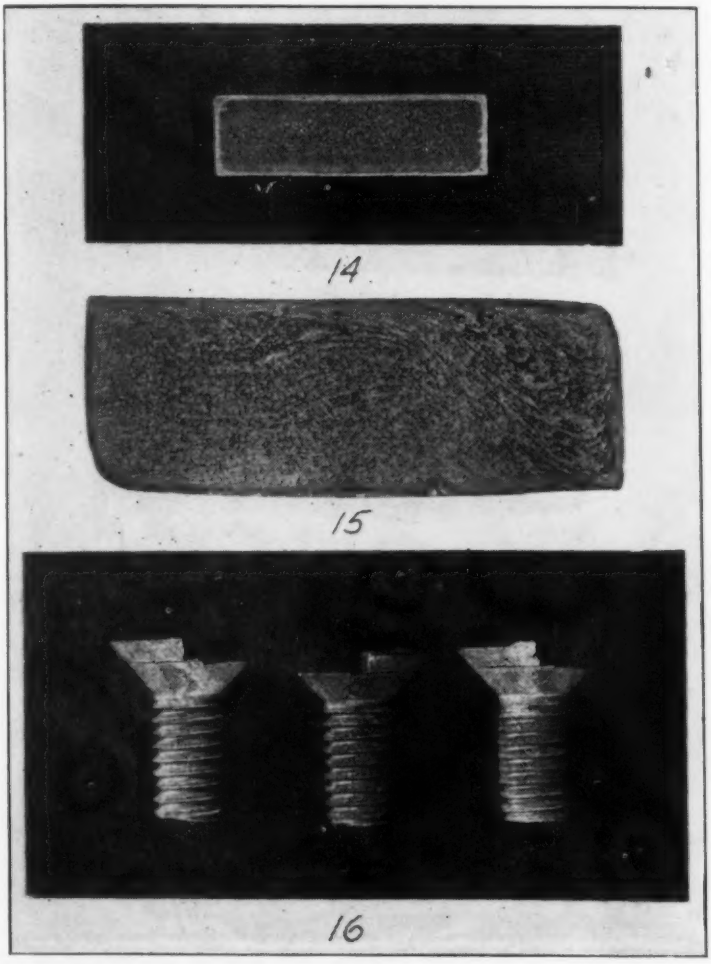


Fig. 14—Decarburized Rim on a Small Bar of High Speed Tool Steel. The Depth of Decarburization is Excessive in this Instance. In Many Cases the Hot Acid Etch will Serve to Show the Amount which Must be Removed from the Surface of Tool Steel to Reach the Normal Metal.

Fig. 15—Characteristic Flow or Strain Lines in a Forged Ring. These Show the Direction of Flow of the Metal and do not Indicate Internal Stresses, as this Specimen was Actually Normalized and Annealed After Forging, Eliminating Any Possibility of Internal Stresses Being Present.

Fig. 16—Cap Screws, the Heads of which Split off in Screwing up with an Electric Screwdriver. The Hot Acid Etch Reveals a Lap on a Level with the Bottom of the Slot in the Head, which Undoubtedly was Responsible for the Weakness of the Heads.

do not indicate internal stresses, as this specimen was actually normalized and annealed after forging, eliminating any possibility of internal stresses being present.

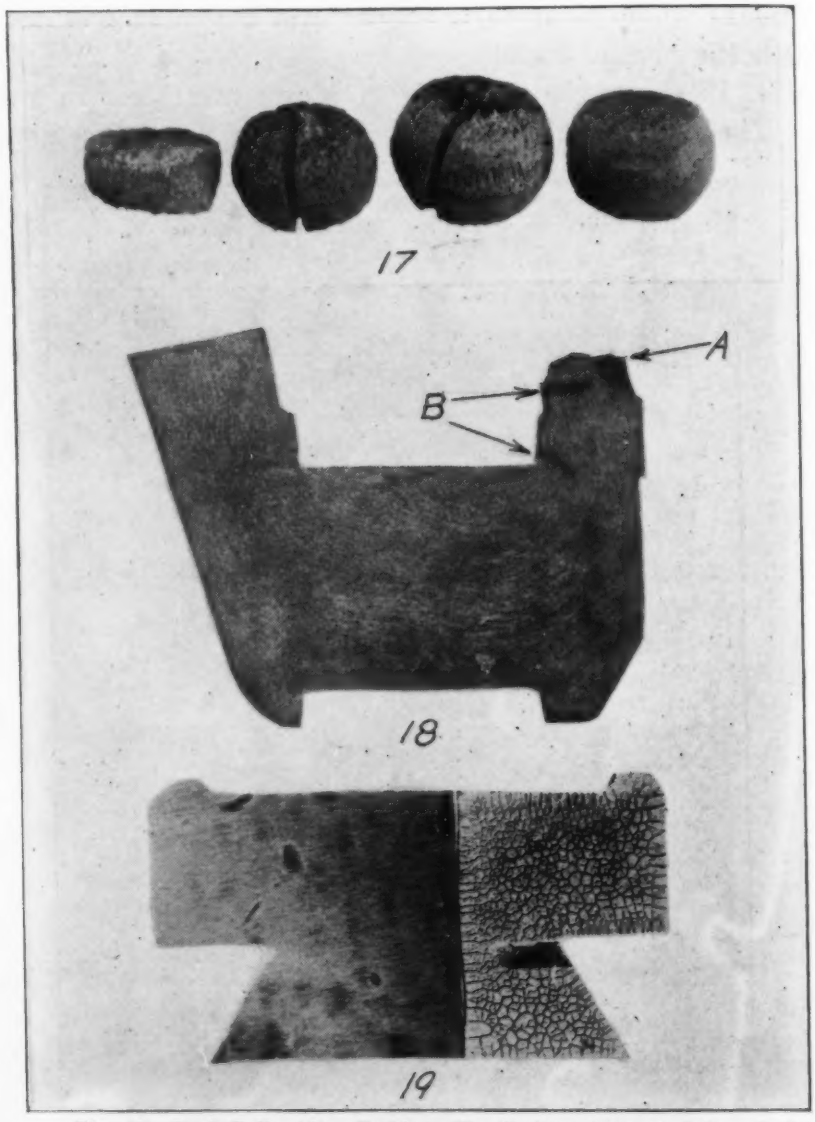


Fig. 17—Steel Balls After Etching. The Etching Characteristics of the Balls Indicate that the First, on the Left, was Turned from Bar Stock, and that the Second and Third were Cold-Headed. The Fourth Ball, on the Right, Appears to have been Hot-Headed.

Fig. 18—Another Example of the Ability of Etching to Disclose Hidden Flaws or Discontinuities in Steel is Shown in Fig. 18. Here, in a Section of a Crank Shaft which Failed in Fatigue, Several Unsuspected Incipient Fatigue Cracks were Found, which Microscopic Examination had Failed to Disclose. "A" Indicates the Actual Fatigue Fracture, "B" Indicates Two Incipient Fatigue Fractures.

Fig. 19—High Speed Steel Form-Cutter which Cracked After Grinding. Etching Indicated that Improper Grinding was Responsible.

Fig. 16 shows several cap screws, the heads of which split off in screwing up with an electric screw-driver. The hot acid etch reveals a lap on a level with the bottom of the slot in the head, which undoubtedly was responsible for the weakness of the heads of these screws.

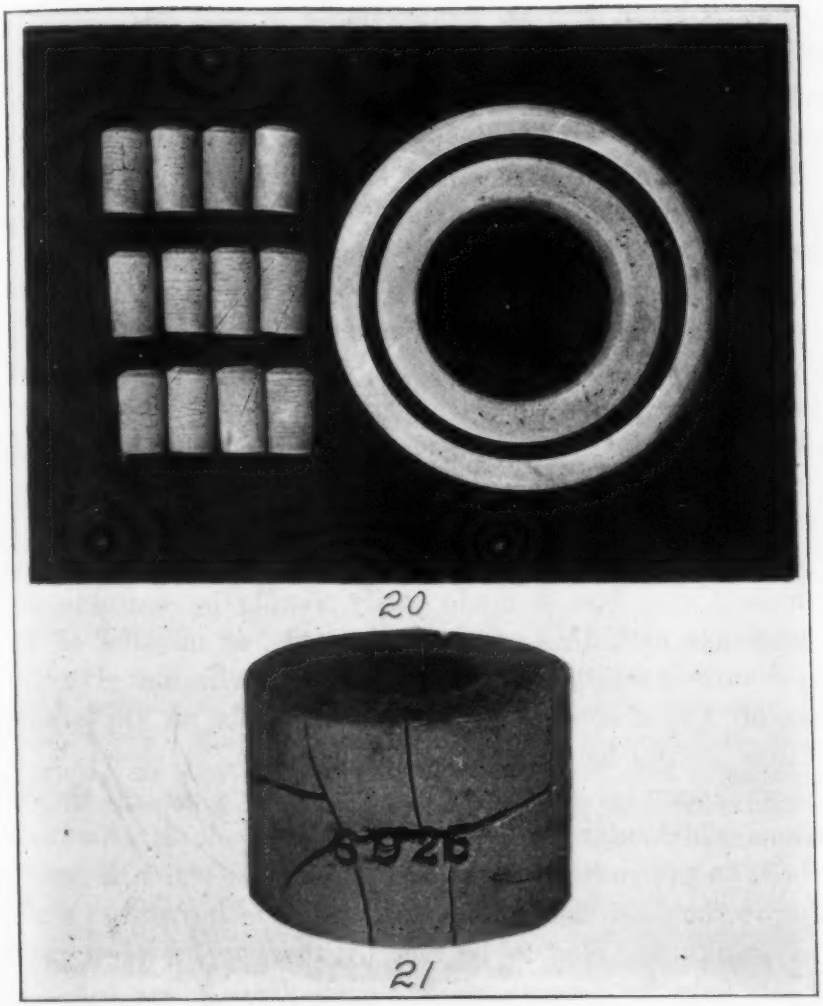


Fig. 20—A Similar Condition is Shown in Fig. 20, Illustrating the Surface Cracks Formed on Roller Bearing Rolls by Careless Grinding. Such Cracks are Invisible to the Unaided Eye, or to the Microscope, but are Readily Made Visible by Etching.

Fig. 21—Effect Produced by Hardening Tool Steel which has been Stamped Deeply. The Stamped Figures Tend to Localize the Internal Stresses Set up in the Hardening Operation, so that Cracking Often Ensues. This is not always Visible to the Unaided Eye, but is made Easily Visible by Suitable Etching. In Connection with this Particular Sample, it May be of Interest to Know that Cracking was Almost Entirely Eliminated by Simply Stoning off the Sharp Crests of the Numerals on the Stamps, to Give Fillets at the Bottom of the Impressions.

Fig. 17 shows four steel balls after etching. The etching characteristics of the balls indicate that the first, on the left, was turned from bar stock, and that the second and third were cold-headed. The fourth ball, on the right, appears to have been hot-headed.

Another example of the ability of etching to disclose hidden flaws or discontinuities in steel is shown in Fig. 18. Here, in

a section of a crank shaft which failed in fatigue, several unsuspected incipient fatigue cracks were found, which microscopic examination had failed to disclose. "A" indicates the actual fatigue fracture, "B" indicates two incipient fatigue fractures.

Fig. 19 shows a high speed steel form-cutter which cracked after grinding. Etching indicated that improper grinding was responsible.

A similar condition is shown in Fig. 20, illustrating the surface cracks formed on roller bearing rolls by careless grinding. Such cracks are invisible to the unaided eye, or to the microscope, but are readily made visible by etching.

In Fig. 21 is shown the effect produced by hardening tool steel which has been stamped deeply. The stamped figures tend to localize the internal stresses set up in the hardening operation, so that cracking often ensues. This is not always visible to the unaided eye, but is made easily visible by suitable etching. In connection with this particular sample, it may be of interest to know that cracking was almost entirely eliminated by simply stoning off the sharp crests of the numerals on the stamps, to give fillets at the bottom of the impressions.

These examples are presented to show some of the hidden conditions which may be exposed by the simple macroscopic etch described, and to indicate some of the fields in which it may prove of value to the steel user. No attempt has been made to set standards of soundness, etc., to be met by the various steel products. These can only be determined for each use, by careful study of the conditions surrounding each individual case. Careful study of the macrostructure of failed or defective parts is of the utmost value to any steel user, in that it provides in the majority of cases conclusive evidence as to relation of the quality of the material to the difficulty, since all failures are not due to material.

Just as no particular type of steel can be truthfully said to be the "best" for all purposes, so also can it be said that no steel is entirely "bad"; no matter how bad it may seem to be, there is always some purpose for which it may be fitted. Careful and reasonable use of the macroscopic examination will often indicate for which service or purpose a given lot of steel may best be fitted. In the interpretation of the etch figures produced, it should be borne in mind that no steel is perfect throughout its entire mass; in fact, soundness and uniformity are merely rel-

ative terms. The degree of soundness and uniformity required, or of unsoundness and non-uniformity that may be permitted, can be determined only by careful consideration of the use to which it is to be put. The steel user should use the same degree of intelligence and discretion in selecting and passing separate lots of steel for a given purpose as he does in the selection of labor. A "foreman" or "superintendent" grading or requirement for a "hunky" job is frequently as wasteful as no requirement at all.

Discussion

CHAIRMAN BIRD: Mr. Curran has in this experience talk given a great many illustrations of the practical utility of the hot acid etch. You have all had experiences contributing towards this subject, and we would all like to hear from you. You have either had your own experiences, or will desire to ask questions pertaining to what Mr. Curran has told you. The paper is open for discussion.

G. F. COMSTOCK: Mr. Curran's paper was very interesting, but it seems to me he was a little too enthusiastic in recommending deep etching. It would be interesting to know what direct evidence he has that this method shows true dirt or oxidized inclusions. It has been our experience that deep etching attacks the sulphide inclusions but it does not seem to me that they should be classified as dirt. The term "dirt" should be confined to oxidized inclusions, such as slag and alumina. It has not been my experience that such inclusions are generally attacked by the acids used for deep etching.

Another objection to deep etching for general use is that it fails to distinguish between a pipe or a crack in the steel and segregated sulphides. They are both attacked and after etching appear as eaten-out places. Sulphur prints, on the other hand, do distinguish between these two defects. Sulphide inclusions cause dark spots while a pipe or crack appears as a white spot on the sulphur print. In my opinion the only way to distinguish or estimate the amount of true dirt in steel is by a microscopic examination of a carefully polished, unetched section.

J. J. CURRAN: Microscopic examination undoubtedly will show the amount of dirt if enough sections are examined. On the other hand, with the hot acid etch one can examine in just one operation a whole cross section and may observe the differences in structure caused by inclusions, whether they are sulphides or oxides. For example, in the case of wrought iron, neither sulphides nor oxides exist alone. A rather complex slag contains both of them, and a lot more beside. The slag streaks show up quite prominently on deep etching. A crack or a pipe, which would be large enough to leave a white streak on a sulphur print, could not fail to be large enough to be visible to the naked eye without deep etching, and I do not see where there could be any question as to its true nature without either deep etching or a sulphur print. One specimen that I showed, (not included in paper),

a supposedly very high grade steel, contained 1.10 per cent carbon or thereabouts on the outside. The center portion, which was relatively unattacked by the etch, although it did have several very deep pits in it, showed a microscopic structure corresponding to a 0.60 per cent carbon steel. A microscopic examination was made after the deep etch. One of the very deep pits which appeared in the central section was exactly in line with a streak of alumina conclusions, which I failed to point out on the slide on the screen. Apparently, therefore, the deep etch will dig out and disclose the alumina inclusions. I believe it will dig out any sort of inclusion; any change in continuity in the metal will show up to some extent in the deep etch. The difference in solubility of two different metals, or the effect of the acid going around an inclusion will be clearly shown on the surface.

DR. J. A. MATHEWS: The practical application of macro etching is rather a new thing and Mr. Curran has referred to the simplicity of the process as compared with exhaustive microscopic studies. He has also mentioned that it does not take nearly so much experience to use it. I am inclined to differ with him in that regard, because I have had a great deal of experience lately in connection with the crimes that are committed in the name of macro etching. I have seen a great many things, none of which were as bad as the lantern slides shown there, in steel rejected as a result of the macro etch. In the first place I find the time of boiling with different laboratories anywhere from 5 minutes to an hour and a half in the same acid which Mr. Curran mentioned. In an hour and a half the steel generally looks like Swiss cheese. Steel, except for a few inclusions, is soluble in hydrochloric acid, if given time enough. It is attacked in a differential way, usually beginning, as Mr. Curran said, at points of chemical differences in chemical concentration of the constituents. However, the action of the acid, depending on the time, magnifies that defect quite as much as your microscope is likely to. They look a great deal worse after the macro etch than they really are in fact. It is like putting a microscope on a more minute defect and enlarging it a hundred diameters or more, depending on the time of your macro etch. If I had known this paper was "macro structure" instead of "micro structure" I would have brought along a few photographs myself, one particularly where a tool that was sent back to us had made an exceptional record. We thought we had better examine it and see what made the tool so good. On macro etching a section we found a dendritic pattern that looked like the linoleum on the kitchen floor. This particular tool had headed over 500,000 one-inch bolts cold and it also showed a few pits exactly at the center where the hardest work was done. I do not think it is something for the inexperienced to handle. It takes quite as much, if not a little more, experience I should say from my experience in connection with it, than the use of the microscope; at least, there are more people who can use the microscope properly than there are who can use macro etching up to date and properly interpret the results.

J. J. CURRAN: I quite agree with Dr. Mathews that any one of our tests can be used improperly. A certain amount of experience is necessary, at least a certain amount of level-headedness and reason. It is possible to be too

exacting in the interpretation of any test. You spoke of a heavy die that showed a dendritic pattern: We had some cases of extremely dendritic material made into dies and followed their record some time ago, and they are going yet. I am not sure that these drawing die tests were really a sensitive test of the material. If it were used for some other purpose it is possible it might cause trouble. So far we have only negative results, but I still have my suspicions about it. We like to test tools that have made exceptional records, either exceptionally good or exceptionally bad, just to run a deep etch test on them to see if we can find out what the effect of structure is. If we have a lot of tools that are showing poor service and they show a uniformly bad macrostructure, whereas the good tools, or tools that are giving good service, show good macrostructures, it is fairly reasonable to say that the macrostructure has something to do with the performance of the tools, and may be used as a test in the selection of material for the purpose.

H. M. BOYLSTON: I would like to join my testimony to that of Dr. Mathews and Dr. Comstock. After an experience of some twenty years in the use of the hot acid etch, I think perhaps the reason that no more has been written or printed on this subject is because of the difficulty of interpreting the results. I like very much to use the hot acid etch, but I always like to confirm it with other tests. If you want a birds-eye view of Buffalo you take a photograph from the air, and take a picture of the city as a whole. But if you wish to find your way about the streets you get a map of the city.

One point that Mr. Curran did not bring out, which I think might have been shown, and that is that the macro-etch often points the way to the selection of proper micro-specimens. I am inclined to agree with Dr. Comstock about the deduction in regard to non-metallic inclusions with macro-etching; the microscope is more valuable.

W. J. MERTEN: The pictures that were shown here, were to a great extent post mortem examinations on failures of materials, and are particularly dangerous to attack from a standpoint that Mr. Curran has attacked them. My own experience teaches me that there are various conditions that might have produced failures which are entirely foreign to impurities of the material. Concentration of stresses from sharp fillets and sharp tool marks is readily conceivable and are much in evidence and it seems to me produce the failures in even the best grade of a steel. I have found that sulphur printing will in a great number of cases bring out the evidence in much clearer fashion and permit judgment of the cause of the failure, it also eliminates, or, rather, prevents some unsatisfactory conclusions which may be arrived at by the deep pitting effect of the hot acids. The evidence that was presented in regard to the grinding cracks is also of a dangerous character. We know that there are strain lines produced by grinding of an extremely hard surface which will give a preferential etching effect and produce a crack by deep etching where a crack has not been previously evident. It appears to me that the danger of drawing wrong conclusions from deep etching is evident.

H. E. SMITH: We try to write careful specifications, then we inspect the material and test it, find it fully meets the specifications, and after awhile we examine it to see why it broke after we first passed it. I would like to

emphasize what has been expressed, that no one test always tells the story, nor does a certain small variation which is found in chemical composition or tensile strength or what not, necessarily indicate causes of failure. We find, as has been said, a sharp fillet, or something of that kind, and if the same piece happens to be a little off in chemical composition, it does not prove that a hundredth of a per cent of phosphorus or sulphur has caused that failure. The macro-etching is useful in locating points for selecting tensile specimens, or other specimens, for further examination. We quite frequently precede it with a sulphur print. It is not a serious matter to get a piece 10 or 12 inches in diameter, smooth enough for sulphur printing, and nitric alcohol etching. By the way, that is a very useful preliminary agent on a 12-inch slice of an axle, or something of that kind. It frequently shows details not brought out in the deep etching. It will show the changed material just below an acetylene weld or a built-up portion, when the piece has not been annealed. That is a frequent cause of fractures. It is quite desirable to examine the unetched surface very carefully first after it has been carefully polished. That will help to distinguish between actual pores and some inclusion which is later etched out by the acid. Often a little magnification, a hand lens, of 10 or 20 diameters, is very useful. No particular illumination is needed.

One thing in regard to wrought iron I would like to mention. Oftentimes someone who has not studied it very much will find steel that is not there. A light spot in etching a piece of wrought iron is not necessarily steel. Much of the iron used commercially is piled material. There is quite a bit of re-worked scrap in the pile, maybe a piece of iron that is quite new, that has not been used over at all, not very much solidified. There may be another piece that is harder and denser, has been piled more than once, and that will show a white or light gray spot. Examination of a great many such spots by taking drillings with a very small drill, determining the composition, shows they are not steel at all, but a little bit harder iron, but they are often called steel in looking merely at the etching. A bar of steel in a pile of iron is usually much whiter than such spots I have mentioned.

B. H. DeLong: An experience which recently occurred to the speaker may be of interest in connection with the investigation of tool failures. A complaint was entered that some rather large and expensive dies had failed in service. Upon investigation it was found that pieces had been cut from these failed dies and deep etched. All the tools showed a common characteristic upon deep etching, and this characteristic had been pronounced the cause of failure. There was some question in the speaker's mind as to whether the condition shown by the deep etching was really responsible, and the question was asked as to whether any dies which had given good service had been similarly deep-etched. The reply was that it was not considered necessary to destroy for investigation purposes tools which had given satisfactory service. The principal involved in this experience is one not uncommonly met in the investigation of steel failures. We do a great deal of investigating to find out why steel parts fail, but do not do enough investigating to find out why they do not fail. I believe that if we could work more broadly along these

lines and learn, not less about parts that do fail, but more about parts that give good service, and thereby secure both positive and negative evidence upon problems of this character, we would add considerabe to the accuracy of our deductions.

H. M. BOYLSTON: I think another etching agent for macro-structural work which has been very much overlooked is the so-called Le Chatelier-Les Moyne agent which I think is described in the A. S. S. T. data sheets. This is a cuprous chloride solution with alcohol and a little water, and containing hydrochloric acid. The specimen is immersed until the copper is plated out on it, then the copper plating is washed off with a piece of cotton immersed in strong ammonia. It often brings out a beautiful structure which is very helpful. Incidentally this reagent will also help to bring out grinding cracks or stresses, (potential grinding cracks), as Mr. Merten has suggested. I do not agree with Mr. Merten that an internal stress is any less dangerous than natural grinding crack, because you cannot tell when it is going to turn into a crack.

I would also like to say that in the ordinary hot acid deep etching test it is sometimes helpful if the piece of etched material is rubbed gently over a piece of fine emery paper for just a few moments to give contrast. It often brings out the "forging lines," very clearly.

W. J. MERTEN: May I add a few words in regard to the examination of materials without any deep etching for segregation and structural irregularities. I do not believe there is anything more simple than to take a bar, put it under a cold shear and shear a piece from it and examine the section for crescent marks or torn folds, which definitely establishes whether the material is sound or not. It is the only thing necessary for that purpose I find always entirely effective in my work. I never failed to detect a segregated section by a crescent mark. A crack produces a sliver which markings are particularly well defined in high carbon steels.

Author's Closure

The discussion of this paper has consisted largely of criticisms of improper uses, technique of application, and interpretation of the hot acid etch. There has been no definite criticism of the method when used for its proper purposes, when the etch was properly applied, or when interpretation was made with a moderate amount of thought and common sense.

The use of the hot acid etch to distinguish between various types of inclusions, as mentioned by Mr. Comstock, is not, and has not, been recommended by the authors. From the steel user's standpoint the type of inclusions present is of no consequence. He is interested only to know when they are present in excessive amounts or when they are seriously segregated. The question of the actual type of the inclusions present is, or should be, of interest to the steel maker, and he is, or should be, acquainted with acceptable methods of identification.

Dr. Mathews has criticized the wide variation he found in etching time and concentration of etching reagents. It is desirable that these conditions be standardized within each plant, and between different plants, as far as

possible. However, it is necessary to vary the time somewhat with different types of material. For instance, wrought irons usually etch satisfactorily in 15 to 30 minutes, whereas structural and tool steels require longer periods.

Objections made by Mr. Merten to post-mortem examinations are groundless, since most of the data on which he bases his own conclusions must have been derived from "post-mortems" of one type or another. The only possible objection which can be made to post-mortems is the improper use of them, that is, where post-mortems are confined to failures alone, and not to successes. This important point has been rightly stressed by several of the speakers in the discussion. The use of post-mortems of failures to illustrate this paper is necessary—if failures never occurred there would be no necessity for any methods of examination and inspection.

As Dr. Mathews has inferred, most of the extremely bad specimens shown are uncommon, nevertheless, they are common enough to cause the steel consumer considerable trouble and expense; and specimens approaching them in degree of undesirability are common and cause trouble in proportion to their numbers and presence in material intended for severely stressed parts. The extremely bad specimens are, therefore, valuable to show structures that are absolutely and unquestionably bad; hence, structures which resemble them and approach them must be viewed with suspicion.

The fact that the hot acid etch has been adopted by several steel mills as a regular routine inspection method on important work indicates that its value as an inspection aid is recognized. While it was first applied only to material going to consumers, using the etch for inspection purposes, and was included only to forestall rejections by these consumers, it was, later, found that its application in the early stages of manufacture eliminated a considerable amount of rejections within the mill during later stages of manufacture.

This fact alone is conclusive proof of the value of the test, and establishes it as a desirable inspection test for the consumer, by which he can allocate incoming material to his various purposes on the basis of the quality shown, or reject it, if necessary.

A STUDY OF DENDRITIC STRUCTURE AND CRYSTAL FORMATION

By Bradley Stoughton and F. J. G. Duck

Abstract

The authors of this paper discuss the formation of dendritic crystals in over-heated high carbon steel and present a study of such crystallization. Comparison is made, both in structure and hardness, with a normal file steel of approximately the same composition.

Evidence is offered, through the inter-crystalline rupture of the over-heated steel, that the amorphous metal hypothesis does not hold when the crystals are large and there are correspondingly large surfaces of cement. Following the same reasoning, it is thought possible that the inter-crystalline rupture of metals at high temperatures is due to the large size of the crystals at those temperatures, as contrasted with their small size at normal temperatures.

It is assumed that the smaller crystals, as well as the inter-lamellar crystals, that occur throughout the eutectoid areas, were formed as a result of the tremendous pressure brought about by the expansion of the material in its change from austenite to pearlite.

Several photomicrographs are presented in support of these conclusions.

It is also stated that the Brinell hardness numbers of any steel vary inversely as the size of its constituent crystals, although no attempt is made to show any relation between them.

WHEN metals crystallize from liquid masses they commonly form in long crystals having a central spine from which lateral branches grow at right angles, these branches again developing secondary spines at angles of 90 degrees. This is brought out in the large columnar crystal illustrated in the TRANSACTIONS of the American Society for Steel Treating, July, 1925, page 33. The interferences, in what Henry M. Howe has called the pine-tree type of crystal, are brought out by the photograph of a surface

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A paper presented before the Cleveland Convention of the Society, September, 1925. Of the authors, Bradley Stoughton is professor of metallurgy and F. J. G. Duck is instructor in metallurgy, Lehigh University, Bethlehem, Pa. The authors are members of the Society.

of cast antimony in Fig. 1 and by the Tschernoff crystal reproduced in the *Metallographist,* Vol. 2, frontispiece.[1]

Rarely can one see the actual structure of such crystals in three dimensions, because our usual method of examination requires cutting and polishing the metal in one plane. In addition to this limitation, metallographic examination reveals only the result of crystal growth under limited circumstances due to complicated interferences. We, therefore, sought an opportunity of examining a crystal growth large enough to be seen with the unaided eye, yet also capable of microscopic examination.

A dendritic growth of steel crystals, produced by casting the metal at a high temperature, afforded an opportunity of studying this type of crystal. The analysis of the material was as follows:

	Per Cent
C	1.28
Mn	0.22
Si	0.22
P	0.010
S	0.011
Ni	0.06
Cr	0.04

The metal was too tender to endure rolling or forging without cracking, and a specimen was, therefore, easily obtained by breaking with an ordinary hammer. The original specimen is illustrated in Fig. 2, which shows the coarse crystallization, as well as the presence of a few blow-holes. In order to show the crystal structure of the fracture, and the large facets of the individual crystals, Fig. 3 is reproduced at a magnification of two diameters. As a comparison with normal steel, we have included Fig. 4 at four diameters, which is a normal flat file steel (as rolled), having the following composition:

	Per Cent
C	1.16
Mn	0.23
P	0.014
S	0.022
Si	0.141

An examination of Fig. 3 indicates that the metal has fractured by the tearing apart of the crystals along the inter-crystalline cement, instead of by normal intra-crystalline rupture through

[1]This crystal has been reproduced in a number of publications, notably H. M. Howe's "Iron and Steel and Other Alloys," and in Dr. Albert Sauveur's "Metallography and Heat Treatment of Steel and Cast Iron."

cleavage planes. In order to show this more clearly, Fig. 5, of the unpolished fracture, was taken at a magnification of thirty diameters. This reveals how the crystals have been torn out bodily

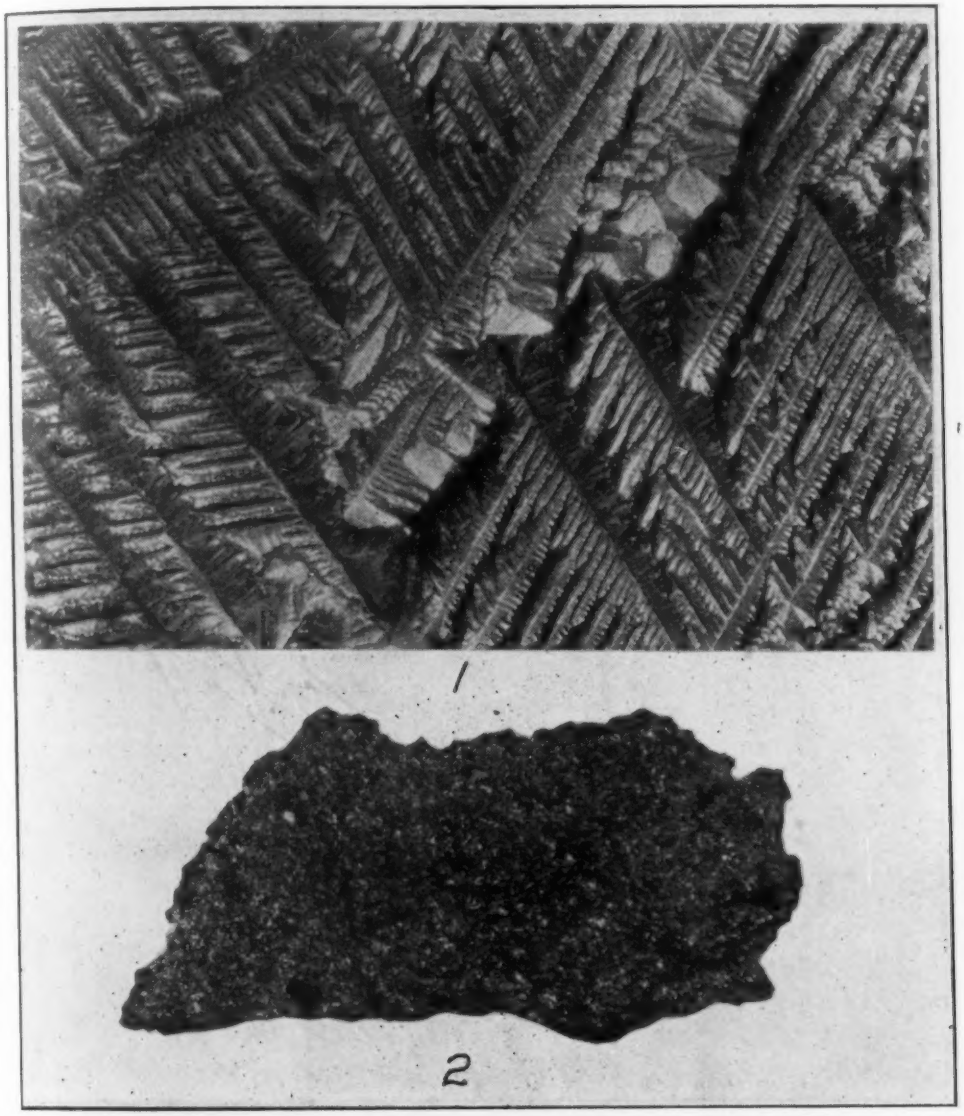


Fig. 1—Photograph of the Surface of Cast Antimony. 4½x. Fig. 2—A Fragment of a Specimen of High Carbon Steel, Showing a Coarse Crystalline Condition. Actual Size.

through the failure of the inter-crystalline cement. Confirmation of this is found in Fig. 6, which is a polished but unetched section of the strained metal showing rupture between the crystals. (NOTE: The relatively greater hardness of the inter-crystalline cementite prevented polishing to a perfect plane surface. Fig. 6

was, therefore, taken with an intentional distortion of the focus, in order that the elevated crystal boundaries would be more evident.)

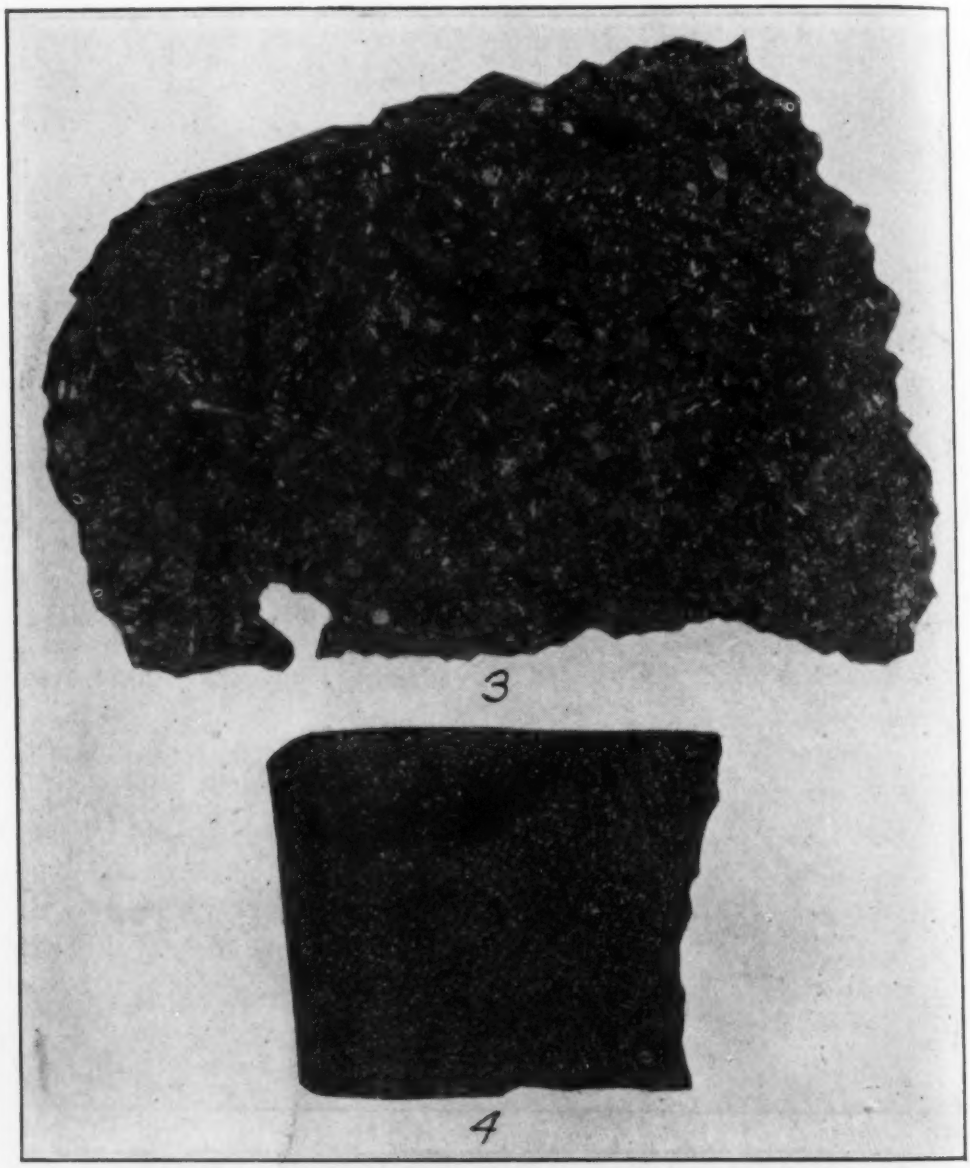


Fig. 3—Photograph of the Specimen Shown in Fig. 2, Magnified 2 Diameters. Fig. 4—Photograph of a Fractured Portion of a Piece of High Carbon File Steel. 4x.

Amorphous Metal Hypothesis

Evidence for the hypothesis of the amorphous, or vitreous, character of the inter-crystalline cement of steel and other metals rests largely on the argument that the cement behaves like a vitreous substance because it is stronger at atmospheric tem-

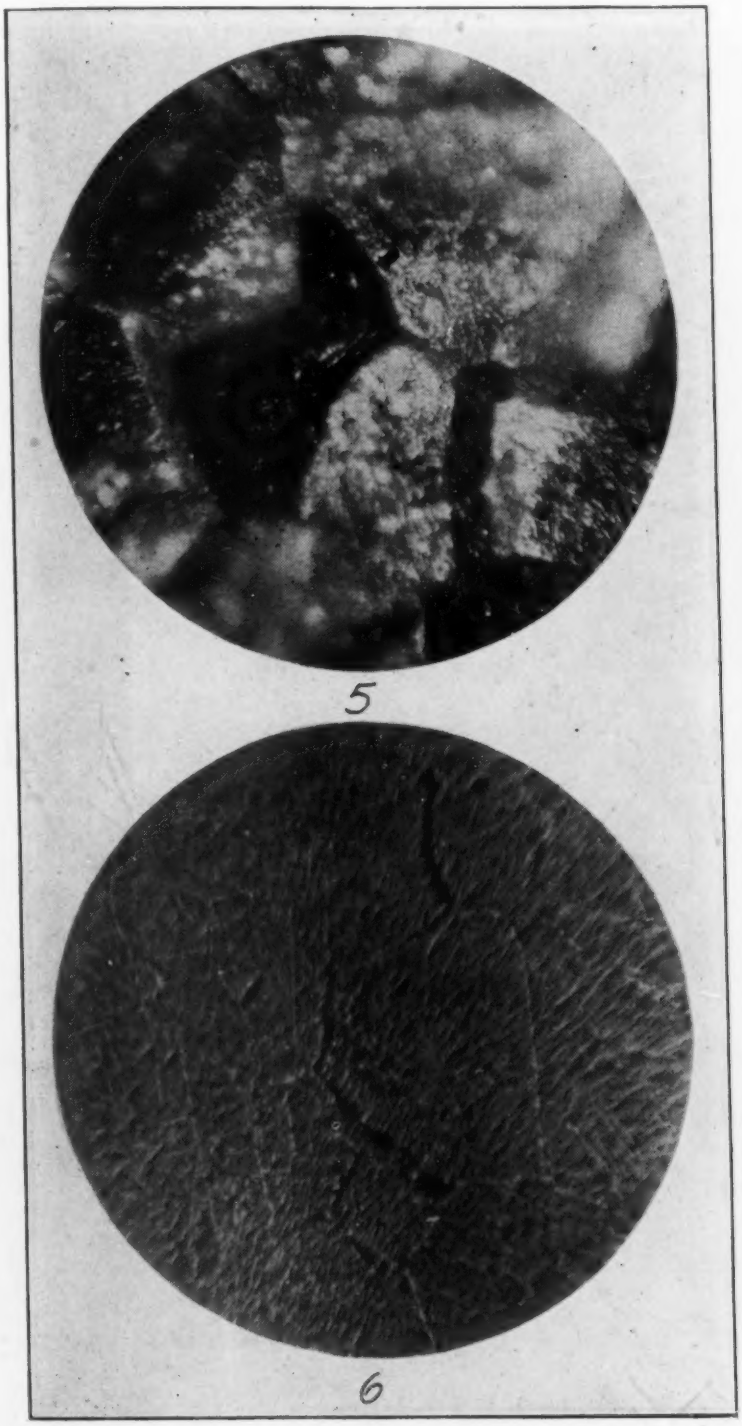


Fig. 5—Photomicrograph of an Unpolished Fracture of the Specimen Shown in Fig. 3, Taken at 30 Diameters. This Photomicrograph Shows How the Crystals Have Been Torn Out Through the Failure of the Intercrystalline Cement.

Fig. 6—Photomicrograph of the Specimen Shown in Fig. 5 After Polishing. The Crack Shows the Rupture of the Metal Between the Crystals. 30x.

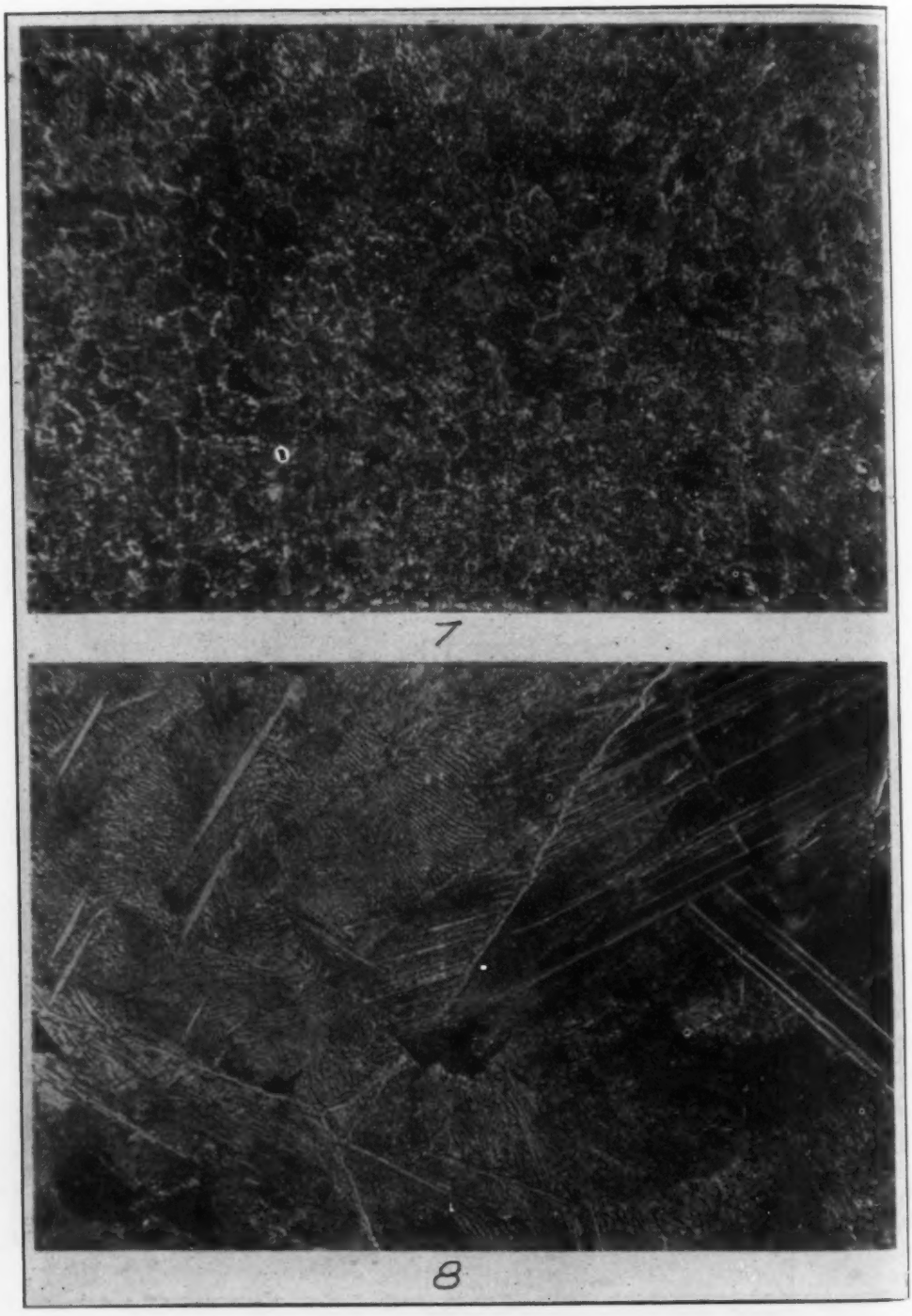


Fig. 7—Photomicrograph of an Etched and Polished File Steel Specimen "as Rolled" and as Shown in Fig. 4. Etched with 2 Per Cent Picric Acid in Alcohol. 190x. Fig. 8—Photomicrograph of a Polished and Etched Specimen of Fig. 2, Showing the Large Spines of Cementite, Especially Along Grain Boundaries. 190x.

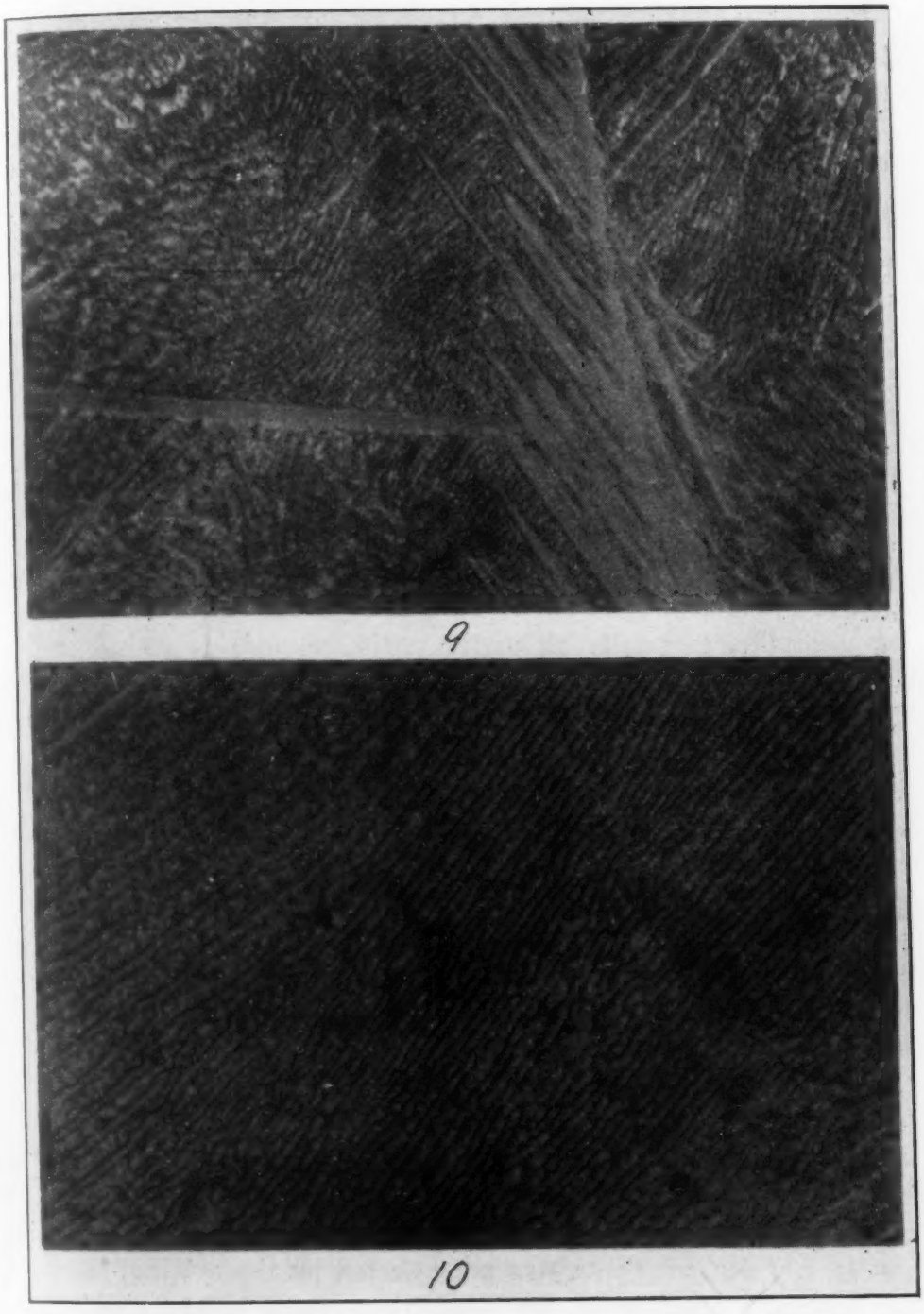


Fig. 9—Photomicrograph of the Same Specimen as Shown in Fig. 8, Magnified 1000 Diameters. Fig. 10—Photomicrograph of the Eutectoid Areas of Fig. 9, at a Magnification of 2400 Diameters.

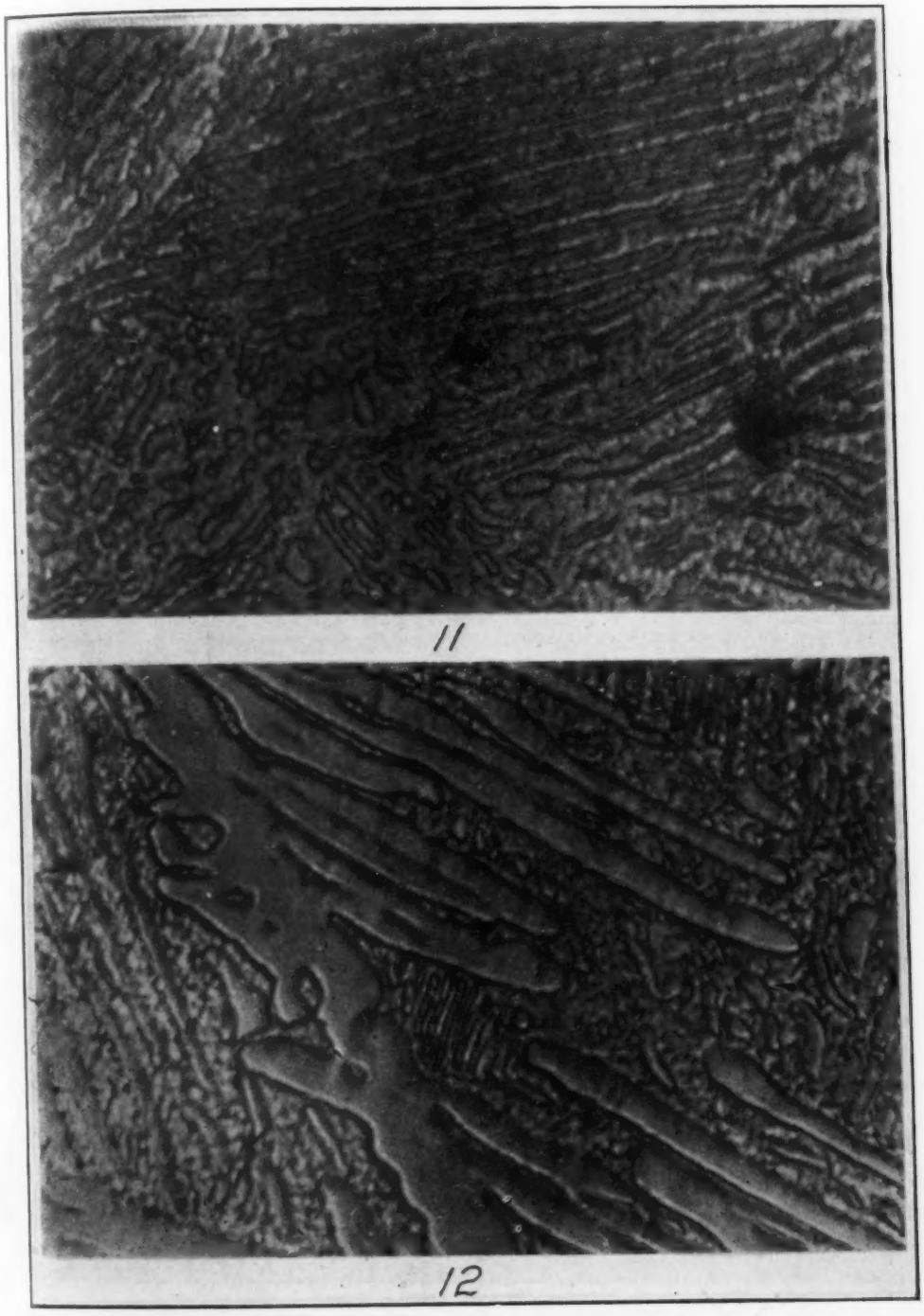
peratures and weaker at a red heat. The proponents of the amorphous metal hypothesis do not claim that their supporting evidence is strong enough to entitle the hypothesis to general acceptance, and many metallurgists, especially in Europe, are averse to accepting the theory because of the universally crystalline structure of metals. The inter-crystalline rupture of the specimen under consideration tends to weaken the amorphous metal hypothesis by indicating that rupture may occur through the inter-crystalline cement when the crystals are large and there are correspondingly large surfaces of cement. Following out this thought, the rupture of metals through the inter-crystalline cement at high temperatures may be due to the large size of the crystals at those temperatures, as contrasted with their small size at atmospheric temperature.

Microscopic Structure

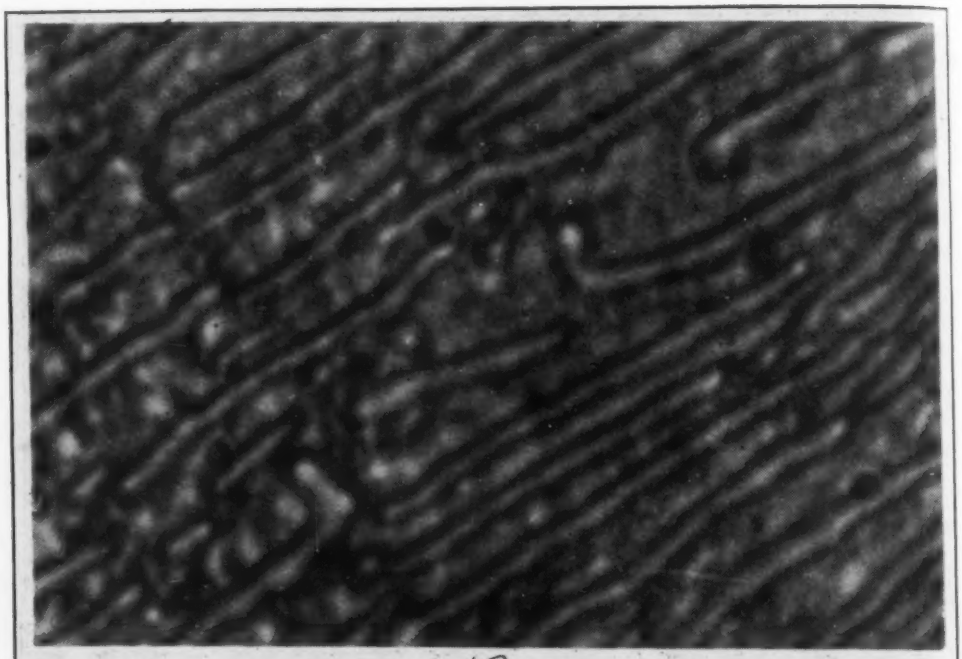
After examining the specimen at magnifications up to thirty diameters, it was etched for 30 seconds with a 2 per cent solution of nitric acid in alcohol. This clearly reveals the large spines of cementite, especially along the grain boundaries, as shown in Fig. 8 at 190 diameters. It is illustrated even more clearly in Fig. 9 at 1000 diameters. If we compare Fig. 8 with Fig. 7, we see that, while the individual crystals of pearlite are much larger in the over-heated steel, nevertheless the crystals of ferrite and cementite, of which the pearlite is composed, are approximately the same size.[2] This is what might be expected, since all the ferrite and cementite were in solid solution when the dendrites crystallized from the liquid mass, but this solid solution had cooled only about 100 degrees when cementite began to separate and migrated to the crystal boundaries of the austenite. Whether these crystal boundaries correspond to the interference between the original spines of the pine-tree growth, is an interesting study which might be made in connection with the crystal illustrated by J. S. G. Primrose in Transactions for July, 1925. It is evident, however, that, in the present instance, the size of the ferrite and cementite crystals which separated from the eutectoid at approximately 700 degrees Cent., was not affected by the original size of the austenite crystals or the dendritic growth in the specimen.

The appearance of the eutectoid at 1000 diameters leads to

[2]The photographs have naturally lost some of their detail in reproduction, and this is particularly true of Figs. 3 and 7. In the latter figure, the crystals of ferrite and cementite, of which the pearlite is composed, are not clearly revealed.



Figs. 11 and 12—Photomicrographs of the Eutectoid Areas of the Coarse Crystalline Steel, Shown in Fig. 8, Taken at a Magnification of 2400 Diameters.



13

14

Figs. 13 and 14—Photomicrographs of the Steel Shown in Fig. 8 Taken at a Magnification of 4800 Diameters. These Photomicrographs are Interesting Since They Show Distortion of the Eutectic Cementite Crystals and a Tendency of the Cementite to Thicken and Fold Back on Itself.

the conclusion that its constituent crystals formed in two degrees of crystal fineness. In order to study this more closely several micrographs were taken at a magnification of 2400 diameters, and three of these are shown in Figs. 10, 11 and 12. Figs. 11 and 12 especially disclose the existence of eutectoid ferrite and cementite crystals of varying sizes and particularly inter-lamellar crystals of irregular size and form. It seems probable that the small crystals, and the inter-lamellar crystals, were formed under the enormous pressure brought about by the expansion of the material in the change from austenite to pearlite. The first crystals to be so formed would be relatively large, but their birth would create a pressure affecting the size and condition of the subsequent crystals.

Examination at 4800 diameters brought out more clearly the structure of the individual crystals within the eutectoid, but attempts to photograph this resulted in a slight "blurring" of the structure because of the long exposure required and the lack of a vibration absorber which has resulted in the very fine micrographs already published in TRANSACTIONS by Francis F. Lucas.[3] Figs. 13 and 14 are interesting, however, since they show distortion of the eutectoid cementite crystals and a tendency of the cementite to thicken and fold back on itself.

The following comparative hardness values of the specimen described and the normal file steel are of interest:

Over-heated steel		File steel
	BRINELL HARDNESS*	
233	Average	268
250	Maximum	272
219	Minimum	257
	SHORE HARDNESS *(Diamond Hammer)*	
16	Average	26
20	Maximum	30
14	Minimum	23
	(Magnifying Hammer)	
35	Average	47
38	Maximum	51
30	Minimum	42

*10 millimeter ball, load of 3000 kilograms held for 30 seconds.

The Brinell hardness numbers of the over-heated steel bear out the observations made on a large number of heat treated specimens, by one of the authors, that the larger the crystals, the lower the Brinell hardness numbers.

[3] "High Power Photomicrography of Metallurgical Specimens," by F. F. Lucas, TRANSACTIONS of the American Society for Steel Treating, Vol. 4, No. 5, 1923, page 611. "Microstructure of Austenite and Martensite," by F. F. Lucas, TRANSACTIONS of the American Society for Steel Treating, Vol. 6, No. 5, 1924, page 669.

Written Discussion:—By Dr. V. N. Krivobok, Carnegie Institute of Technology, Pittsburgh, Pa.

It seems to be evident that the realization of the importance of studying the fundamentals of metallography is gaining ground. Dendritic structure and crystal formation are indeed two important fundamentals and the publication of this additional paper on the subject, especially in view of its merits, must be sincerely welcomed.

The paper as presented is conveniently sub-divided into several separate studies and the discussion of these will be taken up in the same order as they appear in the paper.

A. Amorphous Metal Hypothesis. The information submitted by the authors is very interesting and instructive. Their ultimate contention, worded in their paper as follows: ". . . . rupture may occur through the intercrystalline cement when the crystals are large and there are correspondingly large surfaces of cement . . . ." may, I feel, be questioned. It must be remembered that the production of very large crystals, which in accordance with their formation will be called "primary," takes place only under the conditions most favorable for the undisturbed formation and growth of such primary crystals. But these conditions are just as favorable for a "selective" growth, which invariably results in the forced enrichment by the alloyed elements of those parts of steel that correspond to the fillings between the dendritic branches and also between the dendritic growths or primary crystals; such a condition is the result of the selective solidification of the alloys which form upon solidification so-called solid solutions.

Hence in the inter-crystalline boundaries, especially in the case of steel that displays crystals of abnormal size, we may expect to find also an abnormally high concentration of impurities. In such a case, of course, the authors cannot be justified in drawing any conclusion as to the merits of the amorphous metal hypothesis on the strength of their experiments. The concentration of the impurities as just mentioned would introduce an entirely new aspect. It is well known that the presence of impurities causes intercrystalline fracture regardless of the grain size.

I would like to offer a few experimental data in support of my discussion. An iron silicon alloy containing 2.54 per cent silicon was furnished to me in the form of an ingot made up of unusually large crystals. The fracture of the ingot is shown on the photograph Fig. 1, reduced to one-third its natural size. The crystals could be easily separated from the ingot by sight blows of a hammer. When separated, the crystals show distinctly the traces of oxidation on the crystalline surfaces and in places the indisputable presence of impurities, visible even to the unaided eye. Fig. 2 shows these separate crystals under a magnification of three diameters. A careful polishing of this alloy furnishes an additional confirmation of the stated facts. In Fig. 3 is shown the polished but unetched surface of this alloy. It is easily seen that most of the impurities, the nature of which is not of any importance as far as this discussion is concerned, are located at the boundaries. The photomicrograph was taken with conical illumination.

On the eighth page of their paper the authors make the following statement: ". . . . the rupture of metals through the inter-crystalline cement at high temperatures may be due to the large size of the crystals at those temperatures as contrasted with their small size at atmospheric temperature."

I fancy that this statement may be incorrectly interpreted. In the paper, the authors deal with large crystals which were termed by me (the

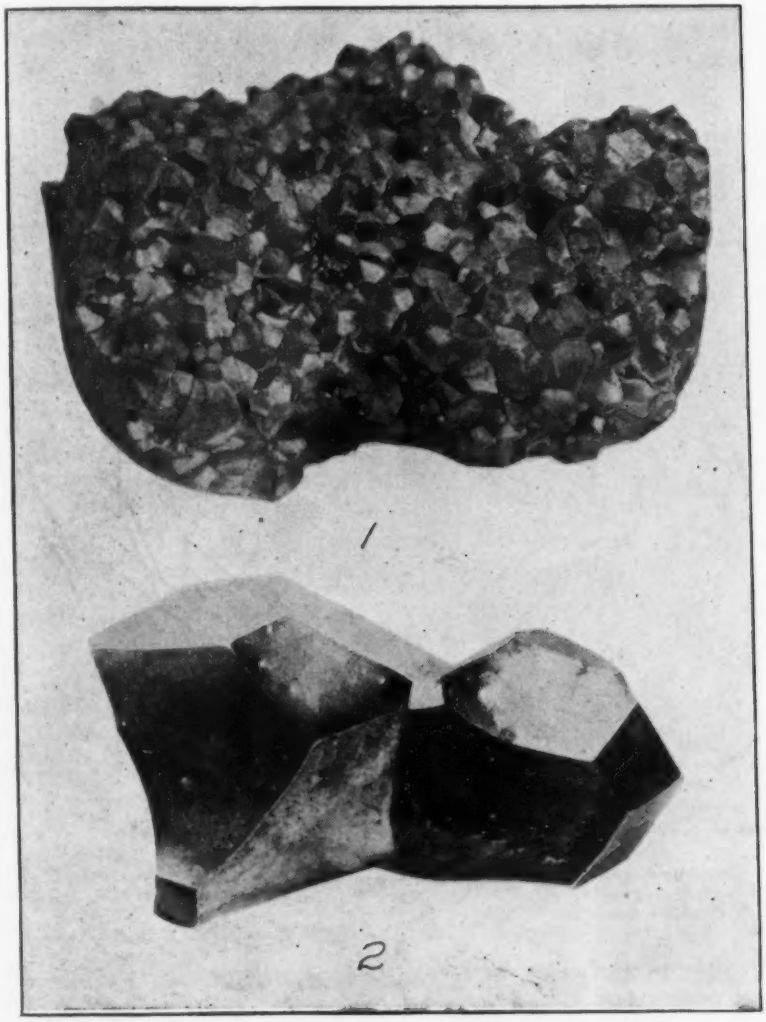


Fig. 1—Fractured Ingot of an Iron-Silicon Alloy Containing 2.54 Per Cent Silicon. ½ Size. Fig. 2—Separate Crystals of This Ingot Enlarged 3 Times.

generally accepted term) "primary crystals" or "dendritic growths." I take for granted that the authors have in mind in the sentence just quoted these primary crystals. The small crystals, at room temperature (mentioned by the authors), which result from re-crystallization in the solid state while the metal cools through the critical range, may not be compared with large primary crystals. The origin. the way of forming, the nature and the dis-

tribution of the internal stresses, of the two are entirely different. It seems to me, consequently, that the statement as quoted is somewhat misleading.

B. Microscopic Structure. I am indeed glad that my observations fully agree with the authors' statement that " . . . . the size of the ferrite

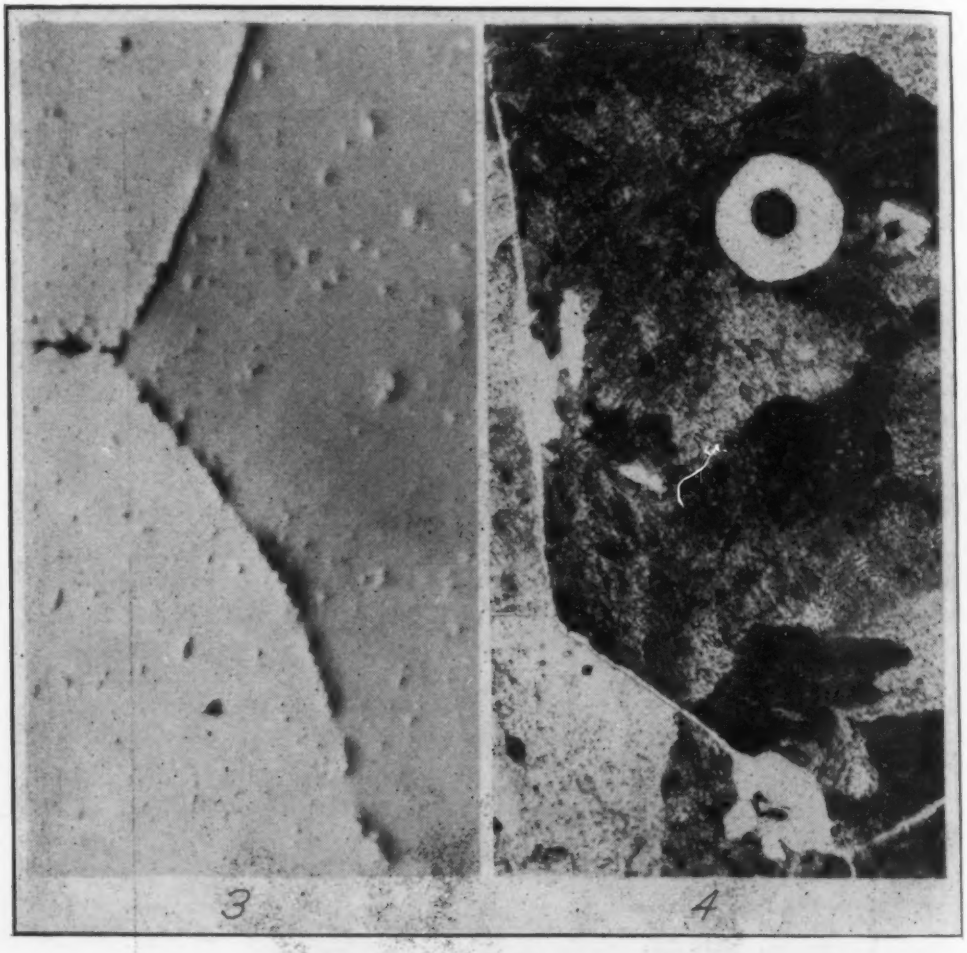


Fig. 3—Photomicrograph of the Polished but Unetched Alloy Shown in Fig. 1. 100X. Conical Illumination. Fig. 4—Photomicrograph of a Specimen of Cast Steel Having Exceptionally Large Primary Grains. 100X.

and cementite crystals which separated from the eutectoid at approximately 700 degrees Cent. was not in any way affected by the original size of the austenitic crystals or the dendritic growth in the specimen."

On the other hand my investigation has shown that the grains of pearlite are only little affected by the original size of the primary crystals. It seems to me that this conclusion is readily substantiated by the theoretical considerations. The formation of pearlitic grains take place below "solidus" in the part of the iron-carbon diagram known as "granulation zone"; i. e., after the primary crystals have been completed. The speed with which the metal is cooled through the granulation zone and in some cases the in-

ternal strain are the factors that are most important in regard to the sizes of the pearlitic grains.

I had a chance to follow this line of investigation suggested by Dr. Sauveur some time ago, and had at my disposal ordinary cast steel of ex-

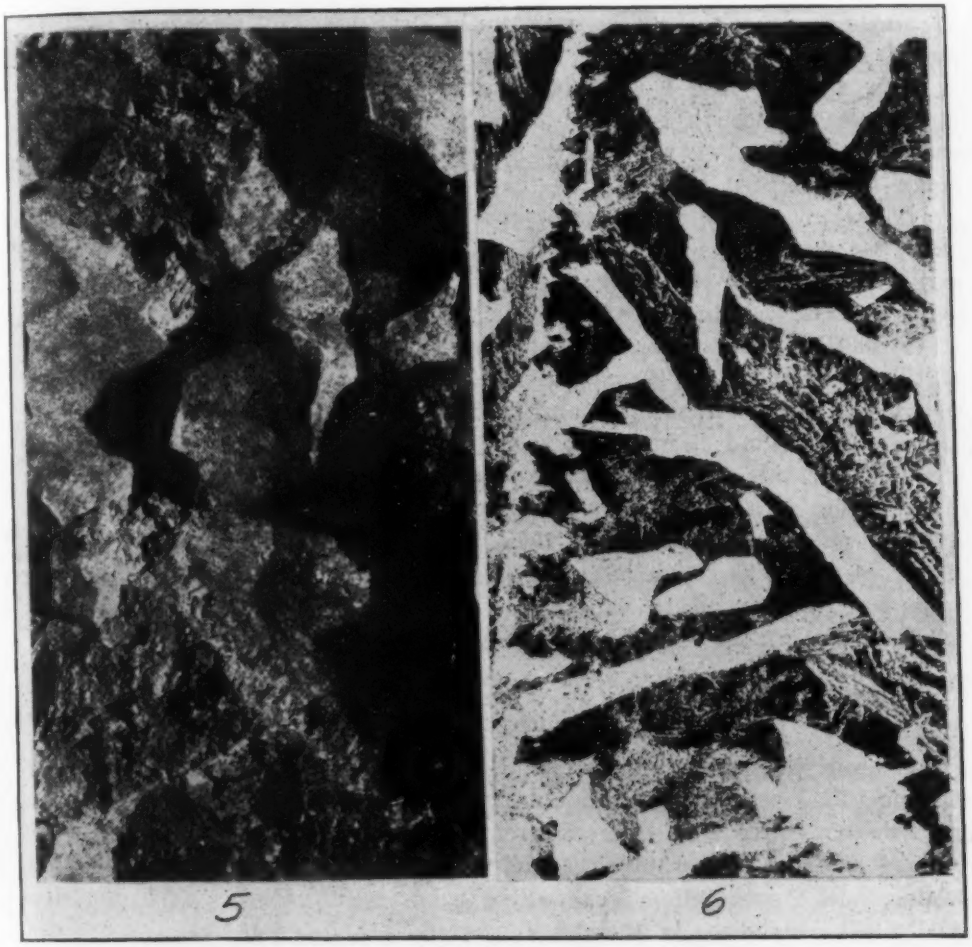


Fig. 5—Photograph of the Fractured Ingot of a 2.50 Per Cent Nickel Steel. Enlarged 3 Times. Fig. 6—Photomicrograph of the Specimen Shown in Fig. 5.

ceptionally large primary grains. The composition of the steel was as follows:

C. 0.63; Mn. 0.79; Ph. 0.04; S. 0.028

The study of the microstructure (Fig. 4) shows that there is no perceptible difference in the appearance of the microstructure. Also, the 2.50 per cent nickel steel, the size of whose crystals can be judged from photographs of Fig. 5, has a structure which does not differ at least under microscopic examination from the same steel of the normal grain size. The microstructure of this large-grained nickel steel is shown in photomicrograph Fig. 6. Accidentally, this photomicrograph shows the boundary between the two adjoining crystals (primary).

It is for the sake of scientific truth only that I take the exception to the authors' statement which, if printed correctly, reads as follows: ". . . . . solid solution had cooled only about 100 degrees Cent. when cementite began to separate and *migrated to the crystal boundaries of the austenite.*" It seems to me that the idea of migration of either cementite as stated by the authors or ferrite (in hypoeutectoid steels) as is often stated in the metallographical literature, is wholly without foundation. It is with apology that I use such a strong term, but I must admit it is impossible for me to conceive the nature of forces that would result in such a migration. I will gladly admit that I am wrong should the evidence be presented to me. But the microscopic evidence, if ever obtained by fellow investigators, has never come to my attention and, on the other hand, the results of my own investigations have failed to furnish the necessary proof for the theory of migration. An attempt may be made to explain this question in a much simpler way if we will admit that some kind of crystalline disarrangement must exist at the grain boundaries. Because of this disarrangement the rejection of the excess constituent at the grain boundaries is much facilitated and that is a reason why we find there most of the pro-eutectoid element. It is out of place here to go into the detailed discussion of this point, but those who are sufficiently interested may find experimental data in the paper, "Secondary Crystallization in Iron-Carbon Alloys," published in the April, 1925, issue of TRANSACTIONS.

It is with sincere pleasure that I wish to thank the authors for their interesting paper and to express the hope that they will continue their valuable research.

Written Discussion:—By Dr. Albert Sauveur, Harvard University, Cambridge, Mass.

I agree with the authors' statement in their interesting paper that when metals crystallize from liquid masses, they commonly form dendrites, but I do not agree with their apparent assumption that these dendrites are retained at room temperature. I believe that these dendrites undergo what Belaiew calls "granulation" on cooling below solidification. Each dendrite, usually of macro-size, is thereby converted into a great many austenitic grains of micro-size, which, on cooling through the critical range, undergo the austenite-pearlite transformation. While the dendrites themselves cease to exist, the dendritic segregation which necessarily accompanies their formation, generally persists and can be revealed by suitable etching treatment. This breaking up of austenitic dendrites into austenitic grains has been explained on several occasions by Belaiew, and I have also attempted a similar elucidation.[1] The temptation would be great to assume that the granulation of the dendrites and the austenite-pearlite transformation both take place on cooling through the critical range were it not for the fact that a similar phenomenon is observed in other metals which are deprived of thermal critical ranges.

Of course when dendrites have been allowed to form like stalactites in

[1]Crystallization of Iron and Its Alloys. TRANSACTIONS American Society for Steel Treating, July, 1923.

the cavities of some castings, or when they are formed in relief on the surfaces of solidifying metals, the outward appearance of the dendrite is retained at room temperature. This is no indication that the dendrites themselves still exist. These apparent dendrites are mere shells filled with austenitic grains above the thermal critical range, and with ferrite and cementite below that range. On sectioning and polishing they will be found to be made up of small polyhedral grains in the case of pure metals, each one with its own crystalline orientation, clearly indicating that the dendrite itself no longer exists as a crystalline unit. The reason, therefore, that dendrites cannot always be detected on polishing a piece of cast metal or of alloy which we know to have solidified as a mass of interlocked dendrites, is not due to the fact that we now can examine it only in two dimensions, but rather to the fact that such dendrites no longer exist.

If, under these conditions, a dendritic structure is revealed, it is merely because of the existence of dendritic segregation permitting a selective action on the part of the etching reagent.

The dendritic structure is essentially a solidification structure, and cannot be produced in the solid metal by any kind of heat treatment. Metals and alloys are not made up of dendrites at room temperature but in many of them dendritic segregation still exists.

In regard to the formation of free cementite in hyper-eutectoid steel as it cools below its solidification temperature to room temperature, this constituent I believe forms first around the boundaries of the existing austenite grains, themselves resulting from the granulation of the austenitic dendrites; and later, if time be given, i. e., if the cooling be sufficiently slow, additional free cementite forms within the grains along the octahedral cleavage planes of the austenitic grains.

There is no evidence from an inspection of Fig. 2 and Fig. 3 of the existence of a dendritic structure. The fracture here exhibited appears to be made up of very coarse cubic grains resulting, I believe, from the granulation of the dendrites, outlined above.

I am in agreement with the authors when they express the belief that the existence of an amorphous phase has not been proven to the satisfaction of all. Admitting its existence, it is not certain that the amorphous phase is more tenacious and harder than the crystalline phase. The greater hardness of a small-grained structure does not need for its explanation the assumption of the existence of amorphous cement surrounding the grains, as it is satisfactorily accounted for on the ground that the smaller the grains, the greater the resistance to slips.

The pertinent question once asked by Professor Howe—why should the contact between a crystalline mass and an amorphous mass of the same metal form a stronger union than contact between two crystalline masses of that metal?—has never, to my knowledge, been answered.

It would be interesting to know what was the casting temperature, and what was the size of the ingot used by the authors.

Written Discussion:—By H. S. Rawdon, Bureau of Standards, Washington, D. C.

One of the conclusions which might be formed from a casual reading

of this paper, though, perhaps, not intended by the authors, is that the fracture of coarse-grained metals, in general, when broken by shock (a hammer blow in the case discussed) is intercrystalline in character rather than transcrystalline, as is generally supposed. The writer believes that it is well established for metals and simple alloys which are brittle enough to break under shock rather than merely to bend, as is often the case in impact tests, the *general* tendency is for the fracture produced to be transcrystalline in its course. This is true for exceptionally large crystals as well as those of more moderate size and numerous citations from the literature in support of this could be given. The introduction of a hard brittle constituent (cementite in this case) as envelopes surrounding the grains might tend to modify this generalization somewhat.

Details concerning the preparation of the "overheated" cast steel used have not been given nor are the photomicrographs sufficiently clear to answer the question whether the inter-crystalline brittleness observed might not be the result of inclusions located between the pre-existing austenite grains. The writer has more than once observed this feature in the structure of steel showing abnormal shock brittleness, and believes that it will account for many such apparently intercrystalline breaks in steel.

It is not clear to the writer why the authors need bring in the very much over worked amorphous-metal hypothesis as related to their observation, at least, not until a more careful study had been made of the grain boundary conditions existing in their material. Relatively few metallurgists accept this hypothesis today in the form in which it was originally advanced by Beilby and Rosenhain. The knowledge of the crystal structure of metals as gained by the X-ray diffraction method has necessitated a very considerable modification of the hypothesis.

Regarding the existence of different degrees of "crystal fineness" in pearlite, the writer would say that it has been his experience that this structural feature can be easily shown without resorting to the use of excessively high magnification. It is a well known fact that the austenite crystal in passing through the gamma to the alpha transformation does not give rise to a pearlite grain uniform throughout its mass so far as the character and orientation of the constituent lamellæ are concerned. A suggestion of this is shown in Fig. 9 of the paper on the left of the conspicuous central grain boundary. Considerable disturbance occurs during the transformation so that in different parts of the pearlite "grain" which resulted from a single austenite grain the character and orientation of the pearlite differ. In the examination of hundreds of rail steels, it has been the writer's experience that this feature of nonuniformity in a pearlite grain is readily observable at a magnification of 500 diameters and in numerous cases the differences are so great that a lower magnification suffices.

It may also be pointed out that the hardness differences reported in the paper are not necessarily to be attributed to grain-size difference alone. The authors apparently are comparing a *cast* steel, of which the rate of cooling after casting has not been given, with a *rolled* file-steel. In order for one to be in a position to make a definite statement concerning the relation between grain size and hardness of these two, he should first of all put them

in a condition so that the comparison can be made by giving the materials identical heat treatment. In a rather comprehensive study made several years ago, of which the writer was joint author (H. S. Rawdon and E. Jimeno-Gil, A Study of the Relation Between the Brinell Hardness and the Grain Size of Annealed Carbon Steels; Bureau of Standards Scientific Paper 397, 1920), it was concluded for the steels investigated that grain size is only a minor factor in determining the Brinell hardness. Reference should be made to the original report for details.

Written Discussion:—By D. J. McAdam, Jr., Naval Engineering Experiment Station, Annapolis, Md.[1]

The fact that the coarse-grained high carbon cast steel specimen shows intercrystalline brittleness is presented as evidence against the amorphous metal hypothesis. The authors assume that if this cast steel had been fine-grained it would not have failed by intercrystalline fracture and they reason that this assumed difference in behavior must be ascribed to difference in the effect of "amorphous cement" in coarse-grained and fine-grained steel.

And yet the authors call attention to the fact that, as shown in Figs. 8 and 9 the grain boundaries are occupied by "large spines of cementite." It is not surprising that metal having a microstructure such as that illustrated in Fig. 8 should show intercrystalline brittleness. The intercrystalline brittleness was probably due to the well known brittleness of cementite rather than to weakness of amorphous cementite.

It is quite possible that if the metal had been cast at lower temperature or were rendered fine-grained by heat treatment it would not show intercrystalline brittleness. Decrease of grain size would mean increase in total area of intercrystalline surface and probable decrease in the thickness of intercrystalline cementite layers. The proportion of intercrystalline cementite might also be decreased. Such decrease in thickness of intercrystalline cementite layers would greatly increase the effective strength of such layers and their ability to withstand distortion.

In order to prove that intercrystalline brittleness may be caused by grain size alone it would be necessary to experiment with metal having only one crystalline phase. Such a metal is difficult to find. I do not contend that it is impossible. In most cases of intercrystalline brittleness occurring in coarse-grained metal a second phase can be found on the grain boundaries. The average thickness of this intercrystalline film would necessarily be about proportional to the average grain diameter.

Moreover, coarse-grained metal sometimes shows intercrystalline brittleness under shock even when it does not show brittleness under slowly applied load. It should be noted that the specimen discussed by the authors was broken by hammer.

In the light of X-ray analysis of crystal structure the "amorphous cement" is merely the submicroscopic layer of distorted space lattice between crystal grains. This layer of distorted lattice probably does not vary greatly in thickness with variation in grain size. It does not seem necessary to assume that at any temperature this layer is weaker than the

[1]Printed by permission of the Secretary of the Navy.

undistorted space lattice. Above the recrystallization temperature plastic deformation under applied stress, either external or internal, takes place not only by slip on cleavage planes but also by distortion and change of orientation of space lattice. This causes continual break up and reorientation of grains and change in position of grain boundaries. Above the recrystallization temperature, even if intracrystalline fracture of a grain occurred, the two fragments due to reorientation under stress would appear no longer as fragments of the same grain but as new complete grain sections.

The variations in the apparent fineness of the pearlite in different grains and the distorted appearance of the cementite lamellæ, described by the authors on the eleventh page of their paper, are probably due chiefly to variation in the orientation of the cementite lamellæ with reference to the plane of the polished surface. As shown by Belaiew[2] variation in the angle at which the polished surface cuts the pearlite lamellæ causes great variation in the apparent fineness of the pearlite. If the cementite lamellæ make a small angle with the polished surface the outcropping plates have the distorted appearance shown in Figs. 11 to 14.

Written Discussion:—By G. F. Comstock, Niagara Falls, N. Y.

Contrary to the impression given by this paper, and especially by the abstract which precedes it, the authors do not offer any evidence which bears upon the amorphous theory. The fracture which they call "intercrystalline" is really inter-cellular, and there is nothing surprising or worthy of note in the fact that in an alloy composed of a brittle material like cementite and a relatively tough material like pearlite, an impact fracture passed through the more brittle material. The presence of the cementite as a network around the pearlite grains of course resulted in the fracture passing around the pearlite rather than through it. This, however, does not mean that the fracture was intercrystalline. The authors do not give any evidence as to whether the fracture occurred through or between the crystals of cementite, and it may, therefore, be assumed that there was nothing extraordinary about it, and that it passed through the cementite crystals in the usual way. The amorphous film which has been assumed by some metallographists to exist between crystals of metals is very much thinner than any network of cementite, such as the authors discuss and illustrate, and cannot be seen with the microscope even at the unreasonably high enlargements which the authors have used for some of their photomicrographs. It is very unfortunate to have such a confusion of ideas regarding the amorphous metal hypothesis appear in this way.

Authors' Reply to the Discussion

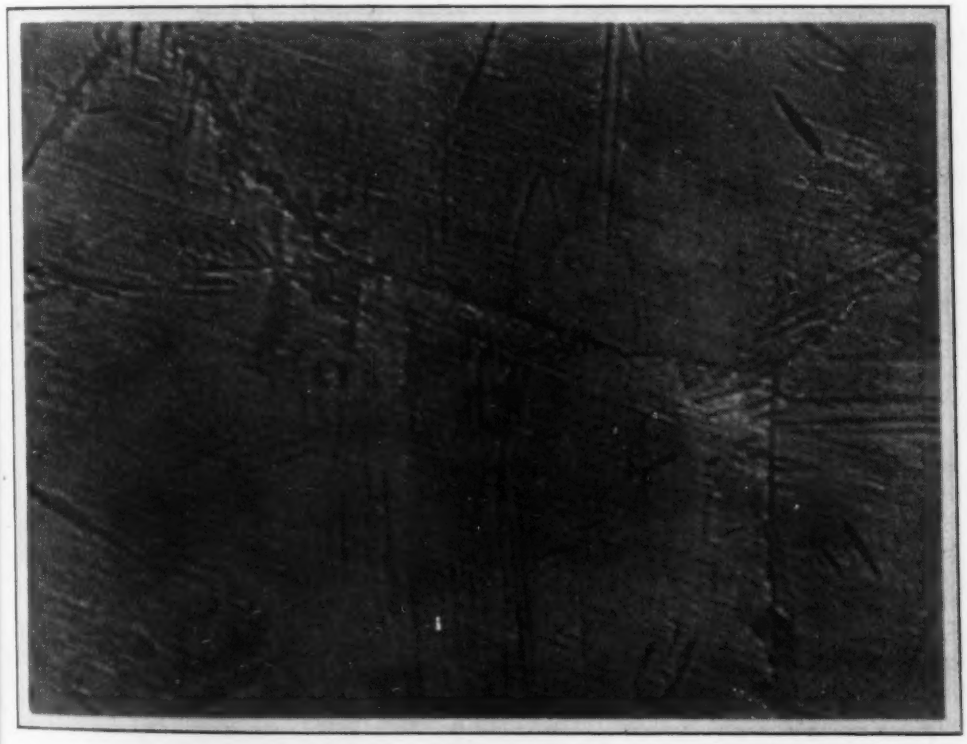
The authors are indeed gratified that their paper was productive of such excellent and extensive discussion (for such was their thought in presenting their observations and hypotheses) but are afraid that some of those who

[2]Colonel N. T. Belaiew, "The Inner Structure of the Pearlite Grain," *Journal*, Iron and Steel Institute, 1922, No. 1, Vol. XV.

participated in the discussion misinterpreted their suppositions and postulations for statements of conclusion and fact. We did not feel that our research was sufficiently extensive to warrant our drawing any conclusions and, therefore, confined ourselves to assumptions only. It is particularly interesting to observe how, in many instances in the preceding discussion, contrary views and opinions on certain theories relating to the amorphous metal hypothesis have been advanced.

We particularly wish to thank Doctors Sauveur and Krivobok for the many interesting and valuable data and conclusions presented by them. Doctor Krivobok's supposition of a "selective crystalline growth" or of a "forced enrichment" is not borne out by our study which showed (in the case at hand) neither of these effects under a wide range of magnifications. However, admitting that Dr. Krivobok's theories of selective growth and forced enrichment are correct, do they not in themselves constitute another argument against the amorphous cement theory?

Dr. Krivobok states that we *may* (not *must*) expect a concentration of impurities at the grain boundaries—but in none of the photomicrographs accompanying our paper are impurities visible at the grain boundaries nor has subsequent study revealed such a concentration under a wide range of magnifications. The accompanying photomicrograph, which covers portions of the same fields shown in Figs. 6 and 8, shows that the grain boundaries are free from concentrated impurities. Further, does not the admission of the concentration of impurities at the grain boundaries (with the consequent



Photomicrograph of the Same Specimen as Shown in Fig. 6. Polished but Unetched. 190 X.

weakening of the intercrystalline "cement") contribute to the weakening of the amorphous metal hypothesis?

Our best answer to the last paragraph but one of Dr. Krivobok's discussion is found in the fifth paragraph of Dr. Sauveur's discussion—with which paragraph we are in entire agreement.

Replying to Doctor Sauveur, we did not intend to imply that dendrites are retained at normal temperatures—we are in complete accord with Dr. Sauveur in his statements in this regard and wish to thank him for his clear explanation of this phenomenon, our discussion (we are afraid) having been somewhat ambiguous.

Although we do admit that the true dendrites do not persist at room temperatures, we contend that large crystals can be retained at normal temperatures, these large crystals filling or making up the "dendritic shells."

The remainder of Doctor Sauveur's discussion is in accordance with our beliefs and observations and, we believe, answers many of the questions raised in the discussions of the other members.

The discussion by Mr. Rawdon is extremely interesting but, we feel, needs no further comment here since the majority of his questions have been answered by our replies to the discussion of our paper by others. His statements in the last paragraph (anent the difficulty, if not the impossibility, of making a direct comparison of a cast with an "as-rolled" steel with respect to hardness) are undoubtedly true, but we cannot agree with him that grain size is only a minor factor in determining the Brinell hardness of steel—our general observations (as stated in our paper) having led us to believe that the Brinell hardness of any given steel varies inversely as the grain size. This belief is in direct accord with results published in the literature—particularly with respect to the brasses and bronzes and other non ferrous metals. Experiments relating to the relation between grain size and Brinell hardness of steels are now under way in our laboratory, from which we hope to obtain definite information in due time.

Referring to the statements made by Dr. McAdam in the first paragraph of his discussion, we remind him that we do not *assume*, but merely *surmise*, (in our paper) that the fracture of the steel in question would have been transcrystalline had the structure thereof been fine-grained. It is a fact, however, (as far as our observations and experience are concerned) that all high carbon (and the majority of medium and low carbon) steels show transcrystalline fracture (and when) the structure is fine-grained.

The other points raised in Dr. McAdam's excellent discussion have already been answered (we believe) either by the discussion of others or by our replies to these discussions.

Certain questions that require for their answers information which is confidential have been aswered by correspondence with those who propounded them.

In conclusion, we wish to again thank the several members for their valued discussions and to repeat that a series of experiments (along the several lines of thought mentioned in our paper) are now in progress in our laboratories. The results of these experiments will, it is hoped, appear in some future issue of TRANSACTIONS.

GRAPHITIZATION AT CONSTANT TEMPERATURE

By H. A. Schwartz

In the June, 1926, issue of Transactions, page 883, appeared the first part of this paper entitled "Graphitization at Constant Temperature". Due to the length of the paper and the appendix, it was necessary that it be printed in two parts. The mathematical derivations (Appendix) referred to in the text, together with the discussion of the paper, are printed on the following pages.

APPENDIX

Mathematical Derivations

Equation 1

Assuming the nomenclature indicated under this equation, and that ρ_c and ρ_m are the densities of the graphite, and of the metal respectively, and r the diameter of the carbon nodules—

It can be shown, assuming that the spheres of carbon are equally distributed, and that the probability of cutting a sphere varies directly as its radius, and inversely as their distance apart, and that the number per unit area will be inversely as their distance apart, that

$$\frac{4}{3}\pi\rho_c r^3 N = \frac{\rho_m M}{100}$$

and $m = 2\,r\,N$. Hence,

$$N = \sqrt{\frac{\pi\rho_c}{.06\rho_m}}\sqrt{\frac{m^3}{M}}$$

$$\sqrt{\frac{\pi\rho_c}{.06\rho_m}}$$

may be calculated from assumed values of ρ_c and ρ_m. Experimentally, $\frac{\rho_c}{\rho_m}$ is found to be between $\frac{1}{3}$ and $\frac{1}{4.4}$. Hence,

$$3.45 < \sqrt{\frac{\pi\rho_c}{.06\rho_m}} < 4.17$$

Say its value is $\frac{1}{4}$.

Then,

$$N^2 = \frac{\pi\rho_c}{.06\rho_m}\frac{m^3}{M}$$

$$MN^2 = \frac{\pi\rho_c}{.06\rho_m}m^3$$

$$= 16\ m^3 \qquad \text{Q. E. D.}$$

Equation (2)

Consider the growth of a sphere of carbon in an indefinitely

A paper presented before the Cleveland Convention of the Society, September, 1925. The author, H. A. Schwartz, member A. S. S. T., is director of research, National Malleable and Steel Castings Co., Cleveland.

large mass of solid solution through which cementite is uniformly scattered.

As soon as a finite sphere of carbon has formed, this sphere will be surrounded by a concentric sphere of cementite free metal. The thickness of this zone will be in a constant ratio to the radius of the carbon nodule which contains the carbon originally of the cementite which has disappeared. Actually the ratio will be about 3:1.

The outer surface of the sphere has a carbon concentration in equilibrium with Fe_3C; the inner in equilibrium with C, i. e., there is a constant difference of concentration between the inner and outer surfaces of the hollow sphere of solid solution. While both the cementite free zones and the carbon nodules are of somewhat irregular outline, they are apparently equiaxed insofar as that term can be here applied, and their average shape may be considered spherical. Below A_1 the structures differ from Fig. 1 only in that ferrite appears in place of the solid solution; the pearlite of the original structure being destroyed. See Fig. 2.

Under such a condition the rate of migration, i. e., the increase in radius of the nodule, will be inversely proportional to the thickness of the sphere or to r, the radius of the carbon nodule. It follows by similar figures, that the mean cross section of the path of conduction is in constant proportion to the area of the carbon nodule, calling t the time and neglecting numerical coefficients throughout.

$$\frac{dr}{dt} = \frac{1}{r}$$

integrating, we find $t = r^2$

or $r = t^{1/2}$

But if c is carbon content

$$c = r^3 = t^{3/2} = t^{1.5}$$

which becomes $c = at^{1.5}$ Q. E. D.

where a is the product of all the numerical coefficients.

Equation (3)

Equation (3) follows from the assumption that if the rate of formation of decomposition products is proportional to the

amount of decomposing material, if k be concentration of agraphitic carbon (the decomposing phase) and t = time

$$\frac{dk}{dt} = bk \quad \text{or} \quad -\frac{dk}{k} = bt$$

The value bt represents the reduction in k in time t; Hence, log k = Constant − bt. But if t = 0, k has its original value k_o, hence

$$\log k = \log k_o - bt \qquad \text{Q. E. D.}$$

Equation (4)

Let k_{gr} and k_{cm} represent the carbon concentration at the inner and outer surfaces of such a hollow sphere as described in the derivation of equation (2); also, let a be its inner and 3a its outer radius, and let k be the carbon concentration at radius r within the sphere, and h^2 the diffusivity of carbon in iron.

Then,

$\frac{dk}{dr}$ is carbon gradient

$4\pi h^2 r^2 \frac{dk}{dr}$ is the carbon conducted across a spherical shell of radius r in unit time. But at equilibrium, when there is neither an accumulation of carbon in the shell, nor a rejection from it, but only conduction, this value is constant for all values of r.

Hence, calling the constant B,

$$k = B\int \frac{dr}{4\pi r^2 h^2} = -\frac{B}{4\pi r h^2} + A$$

The constant B may be made 1 by a suitable selection of the units of measurement, and A is a constant of integration.

But $k = k_{cm}$ when r = 3a, and k_{gr} when r = a. Hence, we can solve for A and dispose of B as indicated. Then,

$$k = \frac{3\,k_{cm} - k_{gr}}{2} - \frac{1}{4\pi r h^2}$$

but the volume of a spherical element of radius r is $4\pi r^2\,dr$ and its total carbon content (mass, not concentration) is $4\pi\rho k\, r^2\, dr$ where ρ is its density.

Substituting for k its value, we find the total weight of the carbon in the sphere to be

$$\int_a^{3a} 4\pi\rho r^2 dr\left(A-\frac{1}{4\pi r h^2}\right) = \rho\left(\int_a^{3a}\frac{4\pi r^3 A}{3} - \int_a^{3a}\frac{r^2}{2h^2} + B'\right)$$

$$= \rho\left(\frac{104\pi a^3}{3}A - \frac{4a^2}{h} + B'\right)$$

where B′ is a constant of integration. The weight of this hollow sphere is

$$\rho\left(\frac{4}{3}\pi(3a)^3 - \frac{4}{3}\pi a^3\right) = \frac{104\pi a^3\rho}{3}$$

Hence, the mean carbon concentration is

$$k_m = \frac{3}{104\pi a^3}\left(\frac{104\pi a^3}{3}A - \frac{4a^2}{h^2} + B^1\right)$$

$$= A - \frac{3}{26\pi a h^2} + B''$$

where $$B'' = \frac{3B'}{104\pi a^3}$$

which becomes by substituting for A and $\frac{1}{ah^2}$, their values in terms of k_{cm} and k_{gr} as calculated from the equation for k, for the values r = a and r = 3a.

$$k_m = \frac{21\,k_{cm} + 5\,k_{gr}}{26} + B''$$

But if $k_{cm} = k_{gr}$, $k_m = k_{cm} = k_{gr}$. Hence, B″ = 0 and

$$k_m = \frac{21\,k_{cm} + 5\,k_{gr}}{26}$$

$$= k_{cm} - \frac{5}{26}(k_{cm} - k_{gr}) \qquad \text{Q. E. D.}$$

Equation (5)

After A. S. Hathaway

"I have found a complete solution of your problem for the mean temperature of a sphere at any time t, given the initial conditions:

1. Its temperature is uniform (v = 1), at time t = 0.
2. Its surface (x = 1), is a non-conductor of heat.
3. Its center (x = 0), is a sink of temperature v = 0."

The following equations are taken from Fourier's "Analytical Theory of Heat," Freeman's translation (pp. 192, 150, y for his x), and are certainly true from y = 0 to y = $\pi/2$:

$$y=\frac{\pi}{2}-\frac{4}{\pi}\sum\frac{\cos(2r+1)y}{(2r+1)^2}\,, \qquad \frac{\pi}{4}=\sum\frac{\sin(2r+1)y}{2r+1}$$

the summation being for r = 0, 1, 2, 3, and so on, indefinitely. Taking $y=\frac{\pi}{2}x$, and $n=(2r+1)\frac{\pi}{2}$ in all equations that follow, which are, therefore, certainly true from x = 0 to x = 1, we have:

$$x=1-2\sum\frac{\cos nx}{n^2}\,, \qquad 1=2\sum\frac{\sin nx}{n}$$

and adding, $$x=2\sum\left(\frac{\sin nx}{n}-\frac{\cos nx}{n^2}\right) \qquad A$$

Therefore, let us take the following solution of the general differential equation (Fourier, p. 269) and note its properties:

$$v=2\sum\left(\frac{\sin nx}{nx}-\frac{\cos nx}{n^2x}\right)e^{-kn^2t} \qquad B$$

1. B satisfies the first condition, v = 1 when t = 0 for all values of x, in consequence of A.

2. B satisfies the second condition, since no heat flows through the surface x = 1 at any time. For, the flow of heat through the surface of radius x (Fourier, p. 92) is, (per unit time, at time t),

$$-4\pi Kx^2\frac{dv}{dx}=-8K\pi\sum\left[x\cos nx+(x-1)\frac{\sin nx}{n}+\frac{\cos nx}{n^2}\right]e^{-kn^2t} \qquad C$$

and this is zero when x = 1, since $\cos n=\cos(2r+1)\frac{\pi}{2}=0$.

It should be noted that the minus sign has been inserted in C so that the flow shall be positive or negative, as it is directed outward or inward.

3. B satisfies the third condition, for taking (from C) the flow at x = 0, we find it to be the negative value $-8K\pi\Sigma e^{-kn^2t}/n^2$, so that the center is a sink of some fixed temperature, which is that to which the temperature v in B reduces after an infinite time, or v = 0, since the exponential factor is zero for t = a, and so v = 0, whatever the value of x.

All conditions being satisfied by the solution B, we find from

it the mean temperature at time t by the integral (Fourier, p. 286).

$$\text{Mean Temp} = 3\int_0^1 x^2 v\, dx = 6\sum_{r=0}^{r=\infty} \frac{e^{-kn^2t}}{n^3}$$

This is readily determined from

$$\int x\cos nx\, dx = x\,\frac{\sin nx}{n} + \frac{\cos nx}{n^2}$$

$$\int x \sin nx\, dx = -x\,\frac{\cos nx}{n} + \frac{\sin nx}{n^2},$$

or respectively, $\sin n/n^2$, $\sin n/n - 1/n^2$, between limits 0, 1.

If $t = 0$

$$6\sum\frac{e^{-kn^2t}}{n^4}$$
$$= 6\sum\frac{1}{n^4}$$

From Fourier, p. 177:

$$\sum_0^\infty \frac{1}{(2r+1)^4} = \frac{\pi^4}{96}$$

Now

$$n = (2r + 1)\,\frac{\pi}{2}$$

Dividing the preceding series by $\frac{\pi}{16}$

$$\sum\frac{1}{n^2} = \frac{\pi}{6} \quad \text{or} \quad 6\sum\frac{1}{n^2} = 1$$

Thus, correctly,

$$1 = 6\,(1 + \frac{1}{16} + \frac{1}{81} \quad \text{etc.})\,\frac{16}{\pi^4}$$

$$= \frac{96}{\pi^4}\,(1 + \frac{1}{16} + \frac{1}{81}\,\text{etc.})$$

The problem of carbon diffusion is exactly parallel and equation (5) is derived from Hathaway's equation by a mere change in terminology, writing

h^2 for Hathaway's k
n for Hathaway's r

and expressing his n in terms of his r, the factor k_o enters to take care of Hathaway's assumption that the original temperature be considered unity.

The series approaches its final values very slowly, indeed.

Its locus for four terms has been plotted, and been shown not to differ visibly from the locus of one million terms. It resembles rather closely a logarithmic curve, so far as it has been followed.

Equation (6)

In an interface reaction the velocity is proportional to the area of contact of the two phases.

Let c be the carbon precipitated, and t the time, r being the radius of the carbon nodule, and a′ an arbitrary constant,

then $$\frac{dc}{dt} = a'r^2$$

but $c = r^3$, hence, $dc = 3r^2dr$ and we have, again neglecting numerical coefficients which may be considered to be all covered by, a new value of a′

$$\frac{3\ r^2\ dr}{dt} = a'r^2$$

or 3 dr = a′dt, hence, $a't = 3r = 3\sqrt[3]{c}$

$$\text{or } c = \left(\frac{a't}{3}\right)^3$$

$$= a\ t^3$$

writing a for $\left(\frac{a'}{3}\right)^3$ Q. E. D.

Written Discussion:—By Anson Hayes, Iowa State College, Ames, Iowa

The first point I wish to take up is the great potential importance of any comparatively rapid qualitative or precision method of measuring graphitization rates.

In 1921 work was taken up by Prof. W. J. Diederichs and myself with a view to shortening the annealing cycle, and, as a result of that work, a paper was presented before the Detroit Convention of this Society in 1922, in which methods of producing malleable iron and intermediate products were presented along with a suggested mechanism for the complete graphitizing action.

The time of forty to forty-five hours to complete graphitization reported there has since been reduced to twenty-two hours on a few samples of white iron and to thirty hours on a considerable number of such irons, all of which were of representative white iron compositions. During this work, we have encountered two samples which graphitized much more slowly.

Now the nearness of approach of the annealing cycle in practice to this period of, say thirty hours, will be determined by the uniformity in rates of graphitization which it is possible to obtain under the conditions that may be realized in the industrial plant.

Two of the variables for which there should be control within as narrow limits as possible are:

(1) *Temperature.*—Qualitative results have already shown that rate of graphitization is extremely sensitive to temperature. Thus in order to arrive at a suitable factor of safety in the length of time for any particular stage of the annealing process, the thing that is needed is a convenient method of as high precision as possible to measure the effects of such temperature variation on graphitization rates. Knowing these effects, steps can then be taken in the way of improvement in annealing oven design to eliminate as far as possible these temperature variations and thus the nearest feasible approach to this minimum annealing cycle can be determined as far as the temperature control is a factor.

(2) *The Chemical Composition of the White Iron.*—No doubt a much nearer approach to the minimum annealing cycle will be possible when the information is available as to the influence on graphitizing rate of a variation of the sulphur content from some three hundredths to about eight hundredths of one per cent, of a variation of the manganese content within the limits that are now considered perfectly permissible for white iron or of a variation of carbon from two to two and six-tenths per cent. It is clear that a systematic investigation of the influence on graphitizing rate of white irons whose composition varies through the limits now considered good practice, would show to what extent the minimum annealing cycle may be approached by using present white iron compositions.

There is, too, the certainty that when the causes and extent of variation in rates of graphitization are known that in some cases it will prove possible to reduce such variations in graphitization rates by comparatively simple methods. Then, again, there is the possibility of revising the present limits of composition of white iron or even of going into compositions outside the present limits, and thus the realization of a shorter cycle and good or even better properties than are now obtained.

In short, the problem of shortening of the annealing cycle as it was taken up in our laboratory has proven to be something as follows:

(1) The determination of the mechanism of graphitization.
(2) In the light of the mechanism to arrive at the minimum annealing cycle in the laboratory.
(3) The determination of all of the causes and extent in variation in rates of graphitization, two classes of which have been suggested above.
(4) The determination of methods of reducing such variation.
(5) Knowing the causes and extent of variation in graphitization rates, the making of such changes in furnace design, in the limits of variation in composition of white iron and other factors, as are found possible, and thus to approach, as nearly as may safely be done, this minimum annealing cycle.

There are also a number of other studies which are necessary but which I will not include here, since they have no bearing upon the present paper.

The fundamental importance of a method such as has been developed by Mr. Schwartz in the pursuance of such a program is evident, for this

program involves an enormous amount of work even with the possibilities of a rather rapid method of measuring rates of graphitization such as has been developed here. Then there is considerable difference of opinion with the present annealing cycle in the way of questions as to what are the limiting quantities of chromium and others of the more unusual elements, and questions as to what are the effects of varying quantities of such elements.

Such a method of measuring rates of graphitization makes possible a definite knowledge of the effects of such elements and offers a possibility of a change in the present policies of utilizing certain kinds of scrap. In fact, it would be difficult to foresee all of the ways in which such a method may contribute to the improvement of the present practice in the manufacture of malleable iron.

I wish to emphasize much more strongly than possibly the author was in a position to do, the importance of this work. In fact, I would have welcomed sometime ago such a method of measuring graphitizing rates, for I am presenting at the Syracuse convention of the American Foundrymen's Association the results of two pieces of work for which it was necessary to develop a method of measuring graphitizing rates. I am sure I am in no way violating my obligation to the program of the American Foundrymen's Association, when I state that after considerable study, the time necessary to absorb the free iron-carbide at 1700 degrees Fahr. was taken as a measure of the graphitizing rate. The use of this method has proven simple and rapid in that it is only necessary to heat a number of pieces of the iron to 1700 degrees Cent. quickly, remove them at intervals, cool rapidly and examine them microscopically.

It should be noted that if the constant "*a*" of equation (2), of Mr. Schwartz, proves to be a satisfactory criterion of graphitizability of a sample of white iron, then it follows that the time necessary to absorb the free iron-carbide at the high temperature should also serve as a satisfactory criterion of the same property.

Since it is very difficult to reconcile the operation of equation (3) in the later stages of the graphitization process with the conclusion III that rates of graphitization at constant temperature are determined by rates of migration, I would consider it better to omit any attempt to do so. I believe that the agreement of the last portion of the graphitization curve with the equation for a unimolecular reaction is an accident. I believe that the last portion of the curve represents the process of reducing the carbon concentration to that of equilibrium with graphite, and also that the portion of the curve preceding this is a result of superimposing the last of the absorption of free iron-carbide on that of equalizing carbon concentration throughout the solid solution.

In regard to conclusion IV that "Either the form in which carbon migrates is always the same" at a definite temperature, or carbon in all its forms migrates at the same rate, I wish to state that work in the laboratory at Iowa State College has recently shown that one of the means of migration of carbon through solid solution during the graphitization process is by means of the gas mixture of carbon monoxide and carbon

dioxide, further, through a knowledge of the degree of metastability of cementite and rates of diffusion of these two gases through the solid solution, it should be possible to calculate rates of migration of carbon.

Another mechanism for migration is through the action of hydrogen and methane. In view of the different rates of diffusion of these various gases, it would indeed be remarkable if all rates of migration are the same. In fact, if we accept conclusion II of this paper, we can show that we have multiplied, by a factor of 3, the rates of migration of carbon in the same white iron and at the same temperature simply by varying the pressure of the gas mixture.

At present two forms in which carbon migrates are definitely known. One through the two reactions.

$[C] + CO_2 \rightarrow 2CO$	in one region, and (where [C] represents any metastable form of carbon)
$2CO \rightarrow C + CO_2$	in some other region, and the
$[C] + 2H_2 \rightarrow CH_4$	in one region, and
$CH_4 \rightarrow C + 2H_2$	in some other region.

Thus the statement of conclusion IV, as it is written, does not hold except, possibly, under the particular conditions under which the experiments of Mr. Schwartz were performed.

Since definite proof of the action of these two gas carriers exists, it is evident that the rate of migration of carbon through the metal has no bearing upon the molecular weight of the unit in which the carbon is held. It is clearly a question as to the permeability of the metal to the two gases involved in the transfer of carbon and to the difference in the stability of carbon in the two regions involved. It should be possible to measure the permeability and the equilibrium constants for the gas mixture in contact with the carbon in the two regions. With such quantitative data, the rate of transfer of carbon can be calculated in case reaction velocities are high as compared to rates of diffusion of the gases.

The recent work of Westgren indicates that the carbon in solid solution in gamma iron does not displace the iron atoms in the face-centered lattice, but that it takes up a position in the face-centered cube.

The fact that the iron-carbon eutectoid in white irons occurs on the same A_{2-3} line with the iron cementite eutectoid, which I hope to present at a later time on this program, and the behavior of these alloys on carburizing and graphitizing, points to one or the other of two conclusions. Either there is only one form in which carbon exists in the solid solution, or if two or more forms exist, they are in equilibrium with each other.

With this evidence in mind, I wish to make the suggestion that the boydenite and austenite which Mr. Schwartz has mentioned, may very well differ only in the number of face-centered cubes which have their interior occupied by a carbon atom.

In regard to the equality of solubility of carbon from cementite and from graphite in gamma iron in the pure iron-carbon system, I feel that the statement that these solubilities are not equal can be made. Experiments by Mr. Evans in our laboratory with pure iron-carbon alloys, of high carbon content, show that all massive iron-carbide can be absorbed in an atmosphere

of carbon monoxide and carbon dioxide in a closed vessel and at a temperature of 800 degrees Cent. in about 100 hours. Further, Dr. Maxwell, of our laboratory, has determined the free energy of the reaction 3Fe(alpha) + C(graphite) $\rightarrow$ Fe_3C, at 700 degrees Cent. and finds it has a value of 2281 calories positive, and Fe_3C is thus metastable at this temperature.

There is abundant evidence that pure iron-carbon alloys graphitize to limiting equilibrium values at temperature above 900 degrees Cent.

Thus we would draw the conclusion that Fe_3C is metastable with respect to graphite and solid solution from 800 to 1100 degrees Cent., and that it is metastable toward alpha iron and graphite below and very near A_1. Thus it seems very probable that it is metastable over this whole temperature range and, consequently, the solubility of carbon from Fe_3C, the metastable phase must be greater than its solubility from graphite.

I desire to modify in one respect the picture which the author has given of the various stages of the graphitizing process. This is that the first graphite that is deposited usually is in contact with free iron carbide. If samples are examined after treatments of only a few minutes, examination shows this to be the case.

If such is the case, this carbon becomes surrounded by solid solution produced in the process of breakdown of the iron-carbide. Following this, the process should be that described by Mr. Schwartz.

In closing, I wish to say a few words on "future developments," in which Mr. Schwartz states that a survey of all the known elements from a literary or experimental viewpoint has recently been completed.

Two specific factors which cause variation in rates of graphitization in addition to those reported in the present paper have been investigated at Iowa State College and, as has been indicated, I hope to present before the American Foundrymen's Association. A number of other investigations along this same line are under way. This work has been done in pursuance of the general program of determining to what extent the annealing cycle may be shortened and in complete ignorance of the program which Mr. Schwartz indicates has already been completed in his laboratory. We are including, of course, in our program not only the problem of identifying the causes of variation, but in addition to that, the determination of means of controlling the variations due to such factors. In the case of one very important cause of variation in graphitization rates, we have already shown that there is a very simple corrective treatment.

I shall be very glad indeed to avoid, as far as possible, any duplication of work that is compatible with the rapid development of our program.

In view of the great amount of valuable information that may be gained by use of a method of measuring graphitization rates, and also in consideration of the great potential benefits that may come to the malleable industry in the way of controlling graphitization rates, I wish to say that I consider the statements which have appeared from time to time in the *Journal* of the American Institute of Mechanical Engineers, and elsewhere, to the effect that no substantial decrease in the annealing cycle is to be expected, as being unwarranted and savoring of fundamentalism. I say this with due consideration of the fact that these statements are being made by men who

have had a long tenure in the technical side of the malleable industry and that they are without question among the great authorities in this field. I feel that to make such a statement is to deny to a considerable extent the great importance of the present paper.

Finally, because of the great importance of the method here given of measuring graphitization rates, and of the quantitative data necessary to developing it, I wish to suggest that sufficient quantities of the experimental data and statements of precautions and methods used to obtain as high precision as possible be made a part of the paper for, without doubt, future work, where use of this method may be made, will be greatly aided by it.

Investigations carried out in this laboratory, and the results of which were included in a report presented before the malleable iron session at the American Foundrymen's Convention, shows that a sample of white iron which contained 0.083 per cent sulphur and 0.26 per cent manganese graphitize at rates something less than one-fourth those of a sample containing 0.034 per cent sulphur and 0.24 per cent manganese. Also, it was shown that in this slow graphitizing sample of white iron that the addition of from 0.15 to 0.22 per cent manganese greatly speeded up the rates of graphitization on cooling from the molten state, upon holding at 1700 degrees Fahr. and during the cooling through the critical range. The work has been repeated three times and the results are undoubtedly dependable for this sample of white iron.

Author's Closure

It is naturally a source of gratification to find that a fellow worker considers the suggested method of measuring graphitizing rates as of outstanding importance.

A reading of the original paper will show that the author agrees entirely with Dr. Hayes, that equation (3) is probably empiric, and represents merely an approach to the locus of Hathaway's equation.

Our experiments do not agree with those of Hayes in showing that carbon nodules always form first in contact with cementite. This, however, is not proof that they do not do so, as may be seen from considerations of the extreme improbability, that the plane under observation should pass through the point of contact of carbon and cementite, even if there be one. The tangency of a cementite and a graphite particle could be observed only accidentally, and occasionally even if its existence were universal.

For the purpose intended, it is immaterial, as Hayes himself shows, where this particle forms, for it soon becomes surrounded by solid solution.

An exception must here be taken to the assertion that the stable and metastable eutectoid lie on the same A_{2-3} line. Dr. Hayes' opinion as to the lack of effect of silicon on this line in the metastable system is not in accord with our understanding of the best supported experimental facts.

It is regrettable that the author's original statement, as to conclusion IV, was capable of misconstruction. From the context, it was thought to be evident that this conclusion applied during the progress of a given graphitization.

The migration of CO_2 and CH_4, as well as many other hydrocarbons, is

well known in connection with case hardening. The author does not wish, in the present connection, to express any opinion as to the form in which carbon migrates. It is to be hoped that further study, now in progress, of the properties of boydenite may permit of conclusions bearing on its composition, and hence, perhaps add to our knowledge of the form in which carbon exists and migrates therein. So far, this information does not exist.

Dr. Hayes' extended discussion of work done, and in progress in his laboratory, seems not to bear sufficiently directly upon the subject of the original paper to entitle the present writer to an expression of opinion.

The major difficulties to be anticipated therein are two:

(1) We do not yet know all the variables which influence graphitizing rates, and hence do not know when all but one have been eliminated in a given experiment.

(2) We do know that the effect of the sum of several variables is not the sum of their effects separately.

In his original summary of his paper, the author advisedly and definitely limited its scope to the scientific aspects of the question, and he must, therefore, now refuse to go beyond those limits into a controversial manufacturing problem. Those interested in the difference of opinion between Dr. Hayes and others, in regard to the possible commercial improvements in annealing, will find the matter sufficiently set out in the *Transactions* of the American Foundrymen's Association, covering the 1924 meeting.

It is still the writer's opinion that there is nothing in his own paper, or in the previous publications of his critic, upon which the works metallurgist can build a material improvement along the lines of a decreased annealing time.

This statement is not to be taken to mean that study of the graphitizing reaction may not uncover other previously unknown variables which may be brought under control to the improvement of the annealing time. A prediction of an increased velocity of graphitization must await the publication of definite information regarding the discovery of such previously unknown variables.

The paper contains no references to the effect of manganese and sulphur upon the graphitizing reaction. The author, therefore, does not understand the relevancy of an allusion to that subject in the discussion.

It is well known that the art invented by Boyden, a century ago, was founded upon the fact that the contents in these elements in American irons differed from those in the relatively ungraphitizable European "white heart" irons. The subject was studied something like fifty years ago by Hammer, and also by A. A. Pope.

The writer did systematic work on the subject to verify the latter's conclusions as early as 1904, and has alluded to these facts elsewhere. ("American Malleable Cast Iron." Penton Publishing Co., 1922.) Other experimenters, doubtless covered the same ground independently. There is, therefore, no possible reason to question the reliability of the facts submitted on this subject by Dr. Hayes. If the latter implies that the information is beyond the knowledge, or contrary to the opinion of the author, he labors under a mis-

apprehension which the present comment is intended to remove.

It is to be hoped that scientific publications have value in clarifying our understanding, and, perhaps, in placing our knowledge upon a more nearly quantitative basis, even if they have opened up no fields which had not been the subject of empiric experimentation by many, quite able individuals, who have practically exhausted their industrial possibilities.

Answering a question from the floor as to the effect of phosphorus on graphitization above and below the temperature at which phosphorus migrates, I am not sure that I can reply to that definitely, because I do not have a perfectly clear picture in mind as to the effect of temperature on the migratory rate of phosphorus. There is a reference in the paper to Fry's work on the subject. In general, I would say the temperature is from 450 degrees Cent. up. I do not believe phosphorus has any effect unless there is enough of it so that steadite is formed. If steadite is formed then it is found that it is a difficult compound to break up. In pictures of the mechanism, given the absence of steadite, phosphorus is negligible, and given the presence of steadite, to all intents and purposes we will retain that forever in any ordinary commercial sense.

RECRYSTALLIZATION TEMPERATURES OF COLD-ROLLED ELECTROLYTIC IRON AND OPEN-HEARTH STEEL STRIP

By John R. Freeman, Jr.

Prefatory Abstract

A study has been made of the respective temperatures of recrystallization of electrolytic iron strip and open-hearth steel strip specially selected for deep drawing operations. The study is a continuation of a previous investigation of the comparative cold rolling characteristics of the two types of material. Tensile tests and microstructural studies were made of strip given different amounts of work by cold rolling and annealed at progressively increasing temperatures. It was found that electrolytic iron underwent a marked recrystallization when annealed for ½ hour at 930 degrees Fahr. (500 degrees Cent.), coincident with a pronounced decrease in tensile strength. Incipient recrystallization was noted in a specimen annealed for 24 hours at 750 degrees Fahr. (400 degrees Cent.). In the open-hearth steel a similar marked recrystallization coincident with a decrease in tensile strength was observed in specimens annealed at 1110 degrees Fahr. (600 degrees Cent.), while no marked change was observed in specimens annealed for ½ hour at 930 degrees Fahr. (500 degrees Cent.).

I. Introduction

In a recent publication[1] data were given on the relative cold rolling characteristics of electrolytic iron and of open-hearth steel strip used for deep drawing operations. The tests showed that electrolytic iron would probably be a somewhat superior material for deep drawing operations. This conclusion was based almost entirely on mechanical tests of the strips as rolled, no studies being made of the annealing temperatures of the two types of material.

It was considered desirable to supplement this work by study-

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[1]Freeman, John R. Jr., France, R. D. Comparative Cold Rolling Tests of Open-hearth Steel Strip (Deep Drawing Stock) and Electrolytic Iron Strip. *Technologic Paper* of the Bureau of Standards No. 288.

The author, John R. Freeman, Jr., is physicist (metallurgy), U. S. Bureau of Standards, Washington, D. C.

ing the relation between the amount of cold work, as measured by reduction of thickness in cold rolling, and the temperature of recrystallization of the two materials. The present investigation is, therefore, an outgrowth of the former one. A brief summary of these rolling tests and treatment of the material is, therefore, included.

In this previous work, representative lots of the two types of material were cold-rolled under as nearly identical conditions as possible and in accordance with both mild and very severe cold rolling practice. Three lots of selected open-hearth hot-rolled steel strip especially adapted for deep-stamping work and two lots of electrolytic iron were used. The electrolytic iron was obtained in tube form approximately 0.070 inch thick. The tubes were slit and cold-rolled to the same thickness as the hot-rolled open-hearth strip, 0.065 inch. Previous to the cold rolling all five lots were annealed at 1400 degees Fahr. (760 degrees Cent.) in the factory in the manner usually employed for sheets and subsequently cleaned by pickling.

II. Work on Annealing Temperatures

(a) Materials

There was available from the cold rolling work a small quantity of lot numbers 1, 4, and 5. Their compositions are given in Table I.

Table I. Composition of Materials Studied

Lot Number	Carbon	Manganese	Phosphorus	Sulphur	Silicon	Copper
	per cent	per cent	per cent	per cent	per cent	per cent
1 (Open-hearth)	0.06	0.39	0.012	0.049	0.01	0.11
4 (Electrolytic)	0.06	0.002	0.010	0.012	0.01	0.06
5 (Electrolytic)	0.05	0.010	0.006	0.009	0.01	0.06

Bureau of Standards analysis.

The samples from rolling series No. 1 (Table II) and No. 3 (Table III) were used for study. In these two rolling series the strip was reduced without intermediate annealing from an original thickness of 64 mils to the minimum obtainable by setting up the rolls as tightly as possible.

Series 1 involved severe rolling, the strip being reduced from

a thickness of 64 mils to 6.5 mils, or almost 90 per cent, in six passes.

Table II. Reduction by Cold Rolling. (Series 1)

Pass Number	Sample Number[a]	Lot Number 1 Thickness in mils	Lot Number 5 Thickness in mils	Approximate reduction per cent
0 (annealed)	0	64	64	0
1	1	36	35	44
2	2	24	24	63
3	3	15[b]	18	70
4	4	12	13	80
5	5	10	10	84
6[c]	6	6.5	6	90

[a]In the text the sample number has been prefixed by the lot number; for example, sample 10 is open-hearth steel (lot 1) which has not been cold-rolled (pass 0); sample 50 is electrolytic iron (lot 5) which has not been cold-rolled (pass 0).

[b]Lot No. 1 was used as a "set up" sample and was passed through rolls twice at this setting.

[c]The strips from pass 5 were doubled and sent through for sixth pass.

In Series 3 the rolls were set up previous to rolling as tightly as possible, and the strip then sent through for two passes with a resultant reduction in thickness of approximately 65 per cent

Table III. Reduction by Cold Rolling (Series 3)

Pass Number	Sample Number[a]	Lot Number 1 Thickness in mils	Lot Number 4 Thickness in mils	Approximate reduction per cent
0	0	64	64	0
1	40	22	23	65
2	41	17	15	73

[a]See note (a), Table No. 2.

after the first pass and 73 per cent after the second pass. No noticeable edge-cracking resulted from severe treatment, although the surfaces appeared slightly torn when examined under a microscope.

(b) Tests Made

One tensile specimen from each sample, taken parallel to the direction of rolling, representing each of the steps in the reduction by cold rolling of both series, together with a specimen of the original material not subjected to these rollings, was annealed for one-half hour at each of the following temperatures: 100, 200, 300, 400, 500, 600 and 700 degrees Cent. Sufficient material was available for only one tensile test representative of each condi-

tion of rolling and annealing. Specimens cut transversely to the direction of rolling could not be obtained. An oil bath was used for the annealing at 100 and 200 degrees Cent. and a fused nitrate bath (1 to 1 mixture of sodium and potassium nitrates) at the higher temperatures. The specimens were placed in the bath when at the desired annealing temperature, held for one-half hour and cooled in air. The tensile properties were determined and the microstructure studied as described below.

(c) Methods of Tests

The tensile tests were made with the same fixtures used in the previous work. These are shown in Fig. 1. The threaded ends of the holders were screwed into the standard tensile testing fixtures which were attached to spherical seats in the heads of the tensile testing machine. This method was used in order to obtain as uniform loading as possible over the entire cross section of the test specimen. The dimensions of the test specimen used are also indicated in Fig. 1. The thickness of the strip was in all cases the thickness of the specimen. A 10,000-pound universal testing machine was used with a $\frac{1}{10}$ poise, giving a 1000-pound total range of capacity.

An attempt was made to determine the Brinell hardness, a micro-Brinell hardness testing machine being used. Reliable results could not be obtained because even under a 6.4-kilogram load with a $\frac{1}{16}$-inch diameter ball, the impression showed on the reverse side of all except the thicker specimens and those in which recrystallization had not occurred.

The ends of the broken tensile test specimens were used for the study of the changes in microstructure, a longitudinal section parallel to the edge of the specimen being polished. Care was taken to polish well below any area that might have been affected by shearing in cutting out the specimen.

III. Results of Tests

(a) Tensile Properties of Open-hearth Steel Strip

The results of the tensile tests of the open-hearth steel strip from rolling series 1 and 3 are given in Figs. 2 and 4 respectively. Values for tests of specimens which received the same amount of cold work but were annealed at different temperatures have been

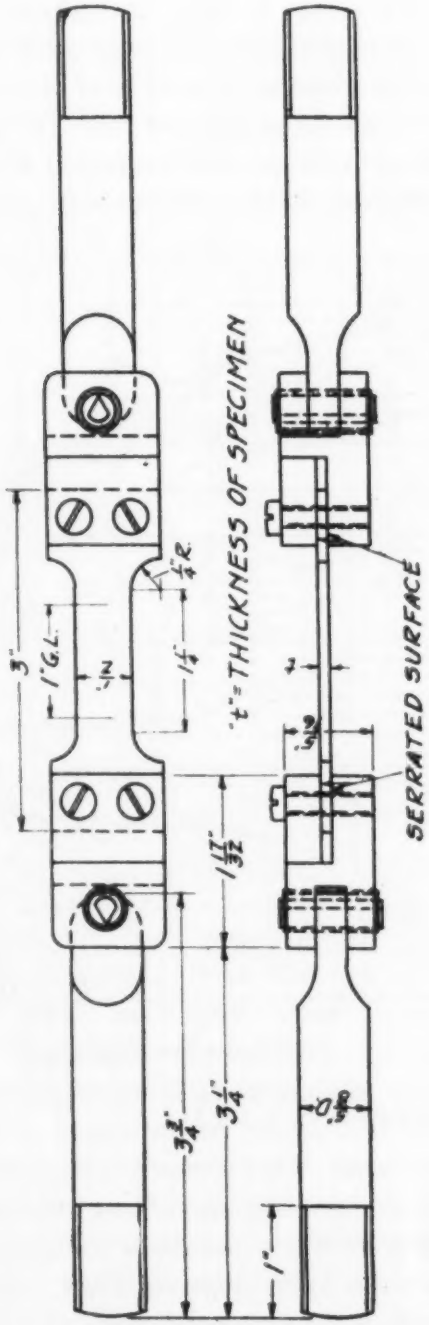


Fig. 1—Fixtures for the Tensile Testing of Thin Sheet Metal

plotted with like symbols. The average curves determined by these like symbols represent the change in tensile strength with increase in annealing temperature for a given amount of cold work.

The curves for series 1 (Fig. 2) indicate that a gradual decrease in tensile strength occurred with increasing annealing temperatures up to 930 degrees Fahr. (500 degrees Cent.); this decrease in strength was decidedly the least for sample No. 11, which received the least reduction. Between 930 and 1110 degres Fahr. (500 and 600 degrees Cent.) there was a marked decrease in

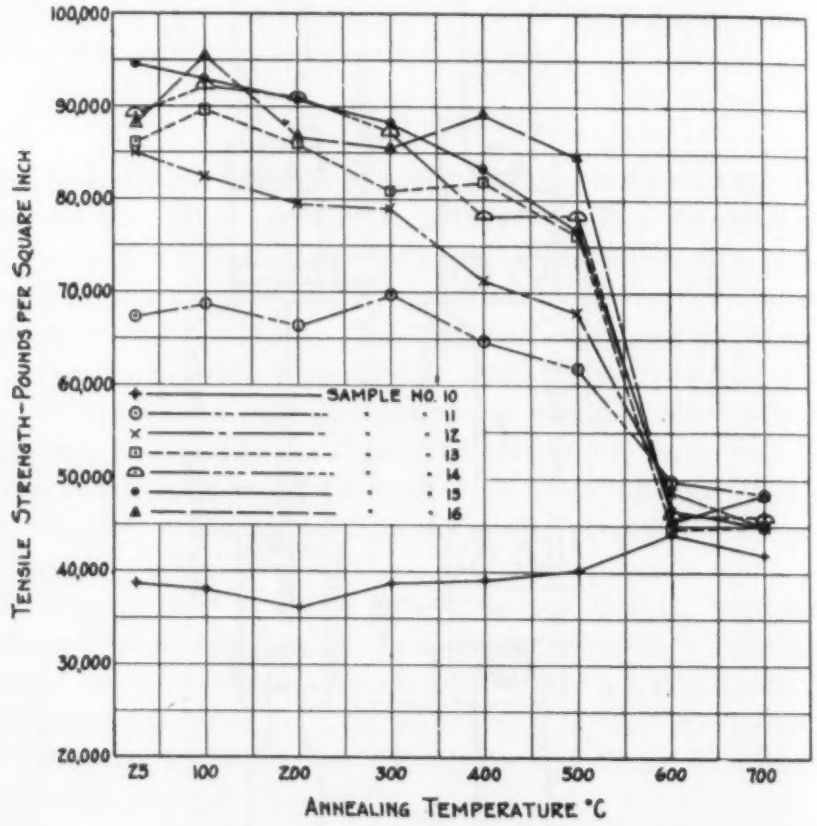


Fig. 2—Effect of Annealing on the Tensile Strength of Cold-rolled Open-hearth Steel (Rolling Series No. 1, Lot 1).

strength. At the latter temperature, all specimens approached the same tensile strength, although those which had received the greater amount of cold work previous to annealing had a somewhat lower value. At 1290 degrees Fahr. (700 degrees Cent.) the tensile strength was approximately the same for all, the amount of cold work previous to annealing apparently having exerted no appreciable influence on the results.

The corresponding curves showing the relation between the tensile strength and annealing temperature for specimens from rolling series No. 3 (Table III) are given in Fig. 4. Few tests

were made but the data confirmed the results of the tests of specimens from rolling series No. 1, in showing a preliminary, but relatively slight, decrease in strength with increase in annealing temperature up to 930 degrees Fahr. (500 degrees Cent.), and between 930 and 1110 degrees Fahr. (500 and 600 degrees Cent.) a very marked decrease in strength.

The results of the elongation measurements for series 1 and

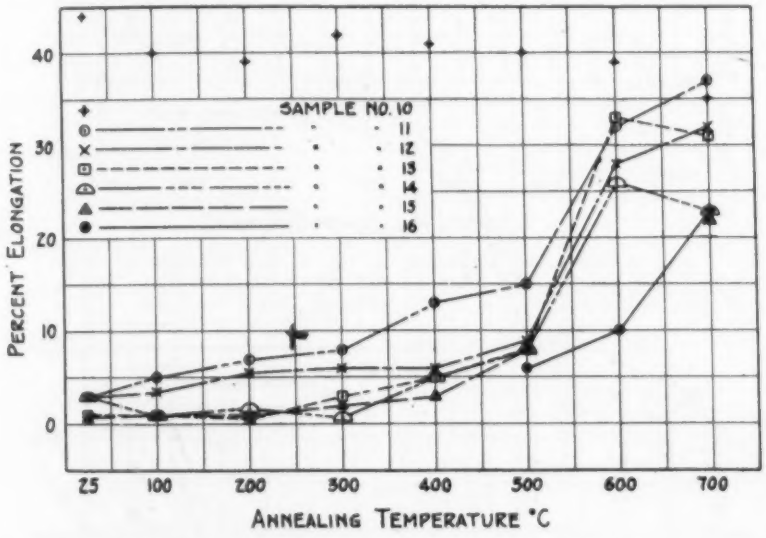


Fig. 3—Effect of Annealing on the Elongation of Cold-rolled Open-hearth Steel (Rolling Series No. 1, Lot 1).

3 are given in Figs. 3 and 5 respectively. They are somewhat erratic, as might be expected from the type of tensile specimen necessarily used. They confirmed, however, quite well the results indicated by the tensile strength in showing a very marked increase in elongation coincident with the decrease in strength within the annealing temperature range of 930 and 1110 degrees Fahr. (500 to 600 degrees Cent.). The fact that the marked decrease in the tensile strengths of cold-worked metal upon annealing is accompanied by an increase in ductility is not new. It has been well established by previous investigators in this field.

Rather pronounced variations were observed in the elongation values of the original materials after being annealed (Figs. 3 and 5). These variations were essentially of the same magnitude, however, as would be found for the initial material before annealing. No consistent variation in the elongation was noted which would suggest a change due to annealing. The elongation of the more severely cold-worked samples was apparently not fully

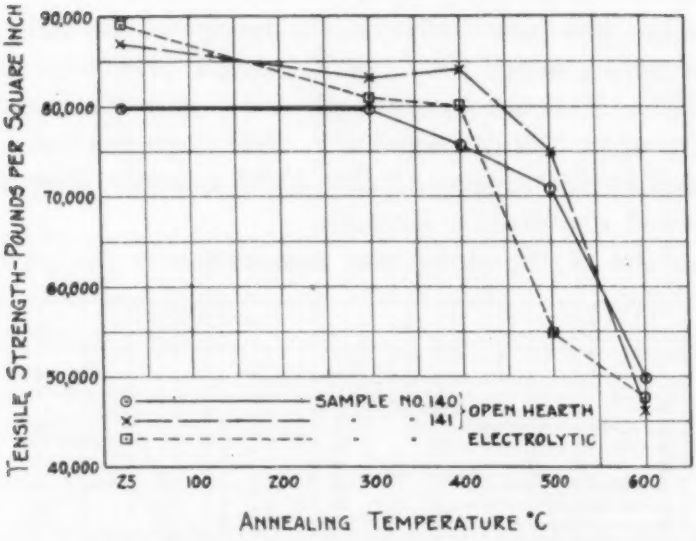


Fig. 4—Effect of Annealing on the Tensile Strength of Cold-rolled Open-hearth Steel (Lot 1) and of Electrolytic Iron (Lot 4, Rolling Series 3).

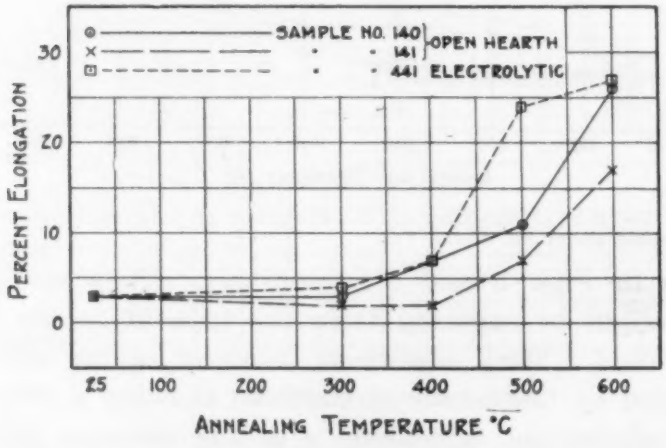


Fig. 5—Effect of Annealing on the Elongation of Open-hearth Steel (Lot 1) and Electrolytic Iron (Lot 4, Rolling Series 3).

restored to its initial value, although the marked difference in elongation between slightly worked and severely worked samples annealed at 1290 degrees Fahr. (700 degrees Cent.) was partially due, no doubt, to the difference in thickness of the specimens.

(b) Tensile Properties of Electrolytic Iron

The tensile strength-annealing temperature curves for the electrolytic iron from rolling series 1 are given in Fig. 6 and those for series 3 in Fig. 4. The general relation of strength to annealing temperature was apparently the same as discussed above, that

is, a decrease in tensile strength with increase in annealing temperature. The electrolytic iron, however, showed a very marked decrease in tensile strength between 750 and 930 degrees Fahr. (400 and 500 degrees Cent.), or about 100 degrees Cent. lower than for the open-hearth steel. The curves for the electrolytic iron show clearly the effect of the amount of previous cold work

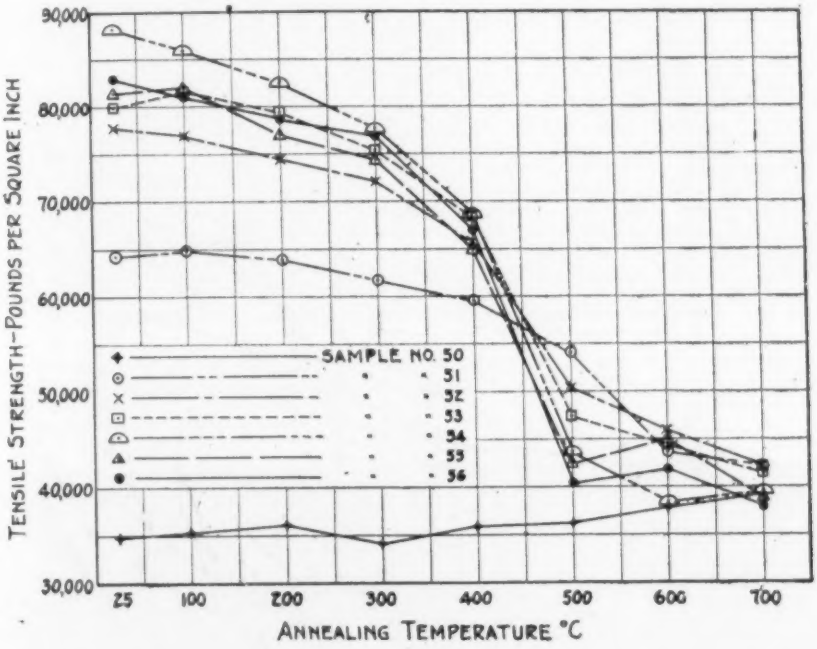


Fig. 6—Effect of Annealing on the Tensile Strength of Cold-rolled Electrolytic Iron (Lot 5, Rolling Series 1).

on the annealing temperature, as is evident in comparing the values for specimens annealed at 750 and 930 degrees Fahr. (400 and 500 degrees Cent.), respectively. At the lower temperature, the tensile values were approximately the same, with the exception of specimen No. 51, which received the least amount of cold work, while at 930 degrees Fahr. (500 degrees Cent.) the tensile strength was lower, the greater the amount of preliminary cold working of the strip. When annealed at 1290 degrees Fahr. (700 degrees Cent.) all specimens, regardless of the amount of cold work previously received, had the same tensile strength.

The results of the elongation measurements of the electrolytic iron specimens from rolling series No. 1 are given in Fig. 7. They showed a gradual increase in elongation with increase in annealing temperature up to 750 degrees Fahr. (400 degrees Cent.), and so confirm the results of the tensile tests (Fig. 6).

Between 750 and 930 degrees Fahr. (400 and 500 degrees Cent.) there was a rapid increase in the elongation coincident with the decrease in tensile strength. As was the case with the open-hearth steel, the elongation of the most severely worked sample of electrolytic iron was apparently not completely restored to the

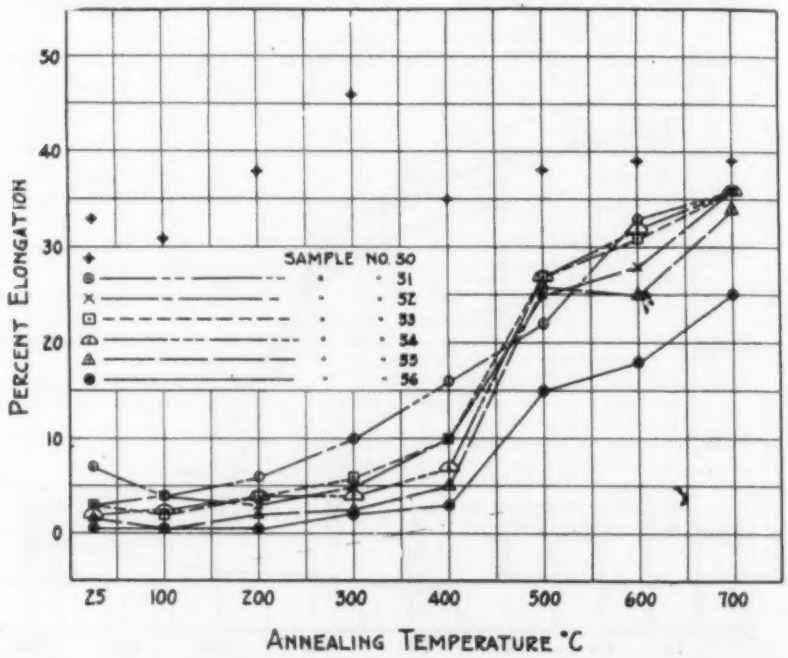


Fig. 7—Effect of Annealing on the Elongation of Cold-rolled Electrolytic Iron (Lot 5, Rolling Series 1).

elongation value of the initial material by annealing at 1290 degrees Fahr. (700 degrees Cent.).

(c) Microstructure

The microstructure of the two materials previous to cold rolling is shown in Fig. 8. The effect of the progressive cold rolling, series 1, on the distortion of the grain structures of the two is evident in Figs. 9 and 10. The structure of the open-hearth strip after annealing at 930 degrees Fahr. (500 degrees Cent.) is shown in Fig. 11. It was evident that no recrystallization had occurred even in the most severely worked material. The small grains shown in the micrograph of specimen No. 16 were free cementite. Annealing this material at 1110 degrees Fahr. (600 degrees Cent.), however, caused a pronounced recrystallization in all specimens (Fig. 12). The effect of the cold rolling was still in evidence, as

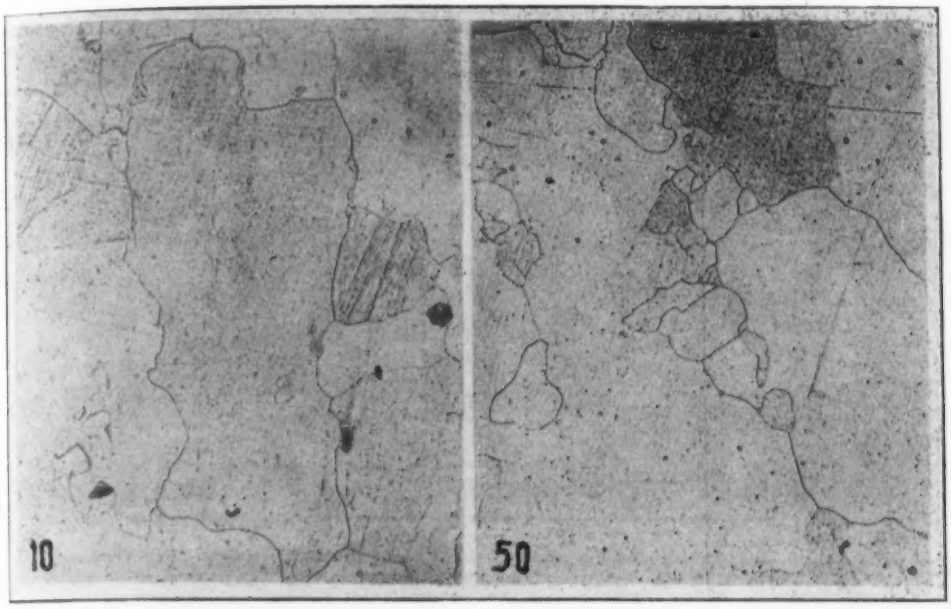


Fig. 8—Microstructure of Open-hearth Steel and Electrolytic Iron Strip Previous to Cold Rolling, 100 x. The Number on Each Photomicrograph in This, and Succeeding Figures, is the Sample Number of the Material. Etching Reagent: 2 Per Cent Nitric Acid in Alcohol. Here and for Succeeding Figures Unless Stated Otherwise.

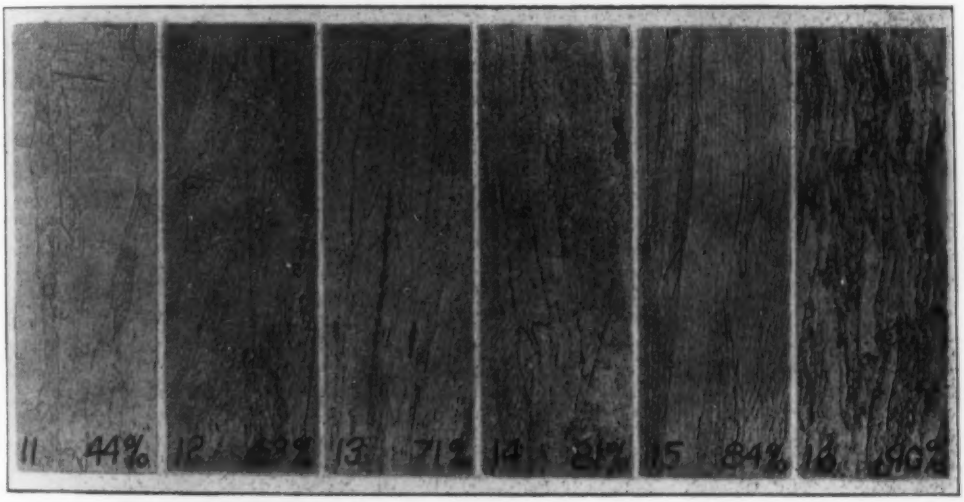


Fig. 9—Effect of Cold Rolling on the Structure of Open-hearth Steel Strip (Series No. 1), 100 x. The Percentage Reduction in Thickness, as Well as the Sample Number, is Given on Each Photomicrograph.

judged by the manner in which the elongated grains produced by the cold work had split up into smaller grains, whose width approximated that of the elongated grains. The effect of the cold work on the refinement of the grain structure was pronounced; the degree of refinement increased with the amount of prelim-

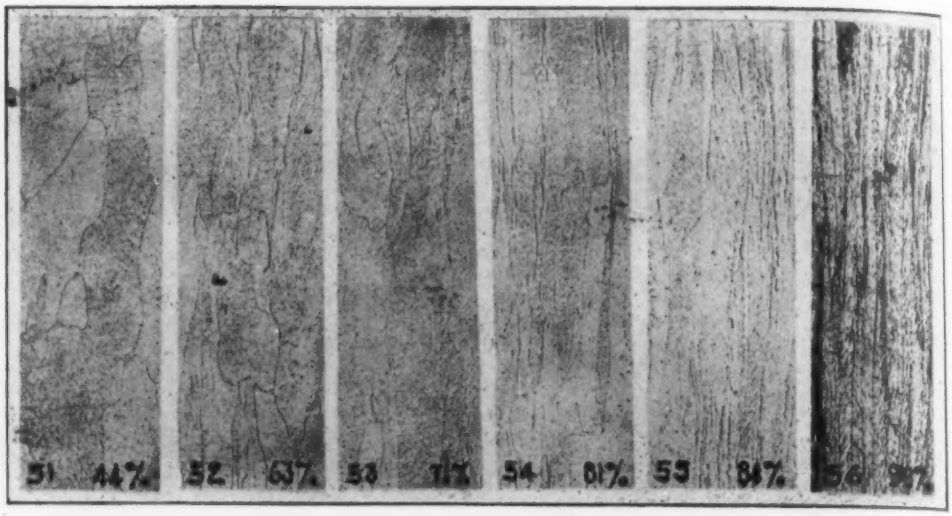


Fig. 10—Effect of Cold Rolling on the Structure of Electrolytic Iron Strip (Series No. 1), 100 x. The Sample Number and the Percentage Reduction in Thickness in Cold Rolling are Given for Each.

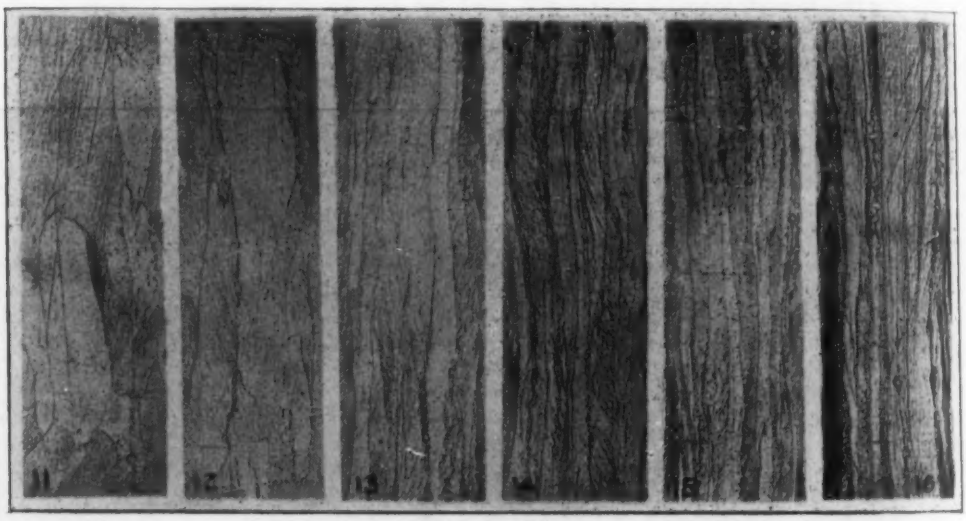


Fig. 11—Structure of Cold-rolled Open-hearth Steel Annealed ½ Hour at 500 Degrees Cent., 100 x. Samples of the Material of Fig. 9 Were Used.

inary cold working. Fig. 13 shows the structure of the same material annealed at 1290 degrees Fahr. (700 degrees Cent.). The recrystallization was quite complete, although evidence of the cold rolling was still present in the "stringing out" of inclusions.

Examination of the microstructure of the cold-rolled electrolytic iron indicated that pronounced, though incomplete, recrystallization occurred between 750 and 930 degrees Fahr. (400 and 500 degrees Cent.). Micrographs representative of the structures

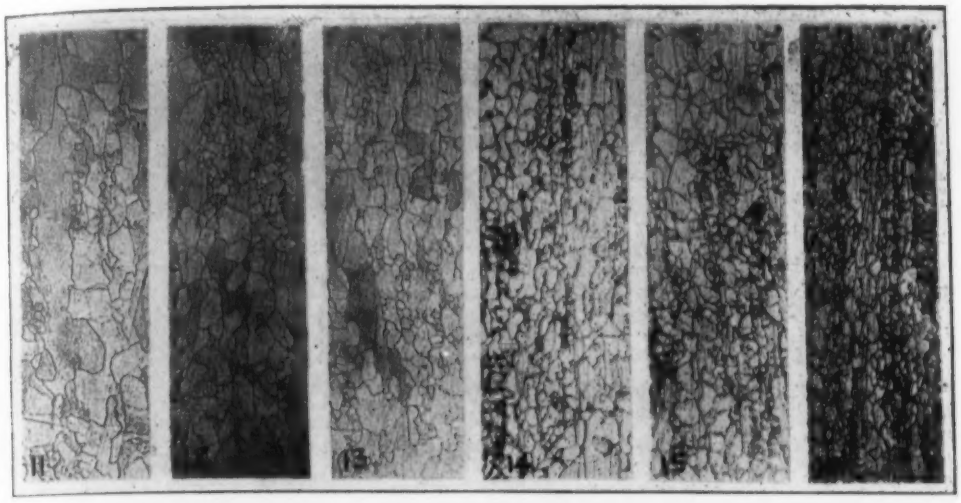


Fig. 12—Structure of Cold-rolled Open-hearth Steel, Annealed ½ Hour at 600 Degrees Cent., 100 x. The Material of Fig. 9 Was Used.

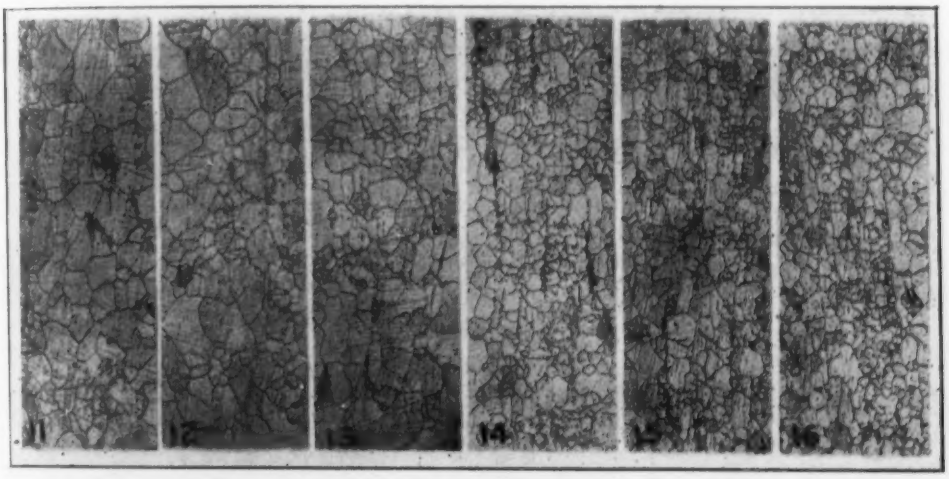


Fig. 13—Structure of Cold-rolled Open-hearth Steel Annealed ½ Hour at 700 Degrees Cent., 100 x. The Material of Fig. 9 Was Used.

are shown in Figs. 14 and 15. Recrystallization was quite complete in all specimens after annealing at 1110 degrees Fahr. (600 degrees Cent.), as Fig. 16 will show.

In order to determine more exactly the temperatures of recrystallization of the electrolytic iron, specimens from the same samples (Rolling series No. 1) were annealed for one-half hour at 840 degrees Fahr. (450 degrees Cent.). The respective microstructures of the specimens were found to be the same as that of this material when annealed at 750 degrees Fahr. (400 degrees Cent.) (Fig. 14). No recrystallization was apparent when exam-

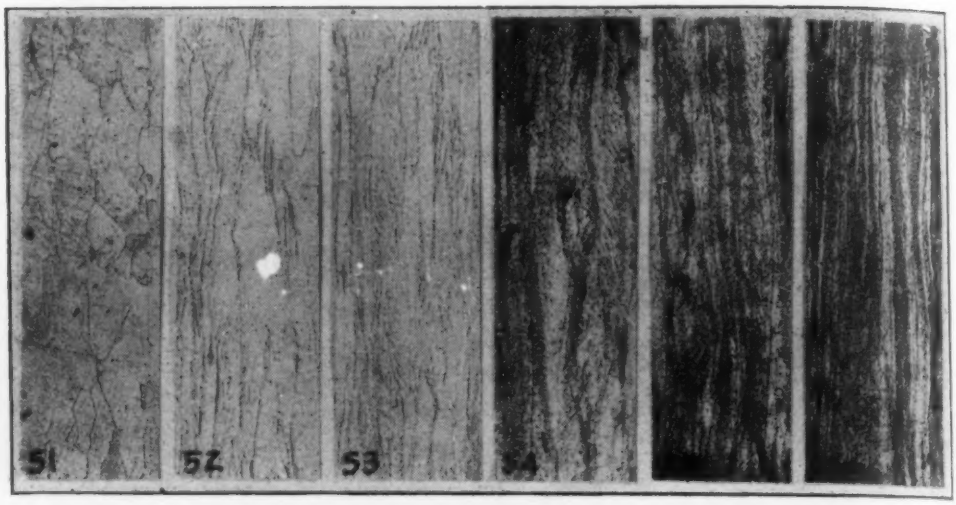


Fig. 14—Structure of Cold-rolled Electrolytic Iron Annealed ½ Hour at 400 Degrees Cent., 100 x. The Material of Fig. 10 Was Used.

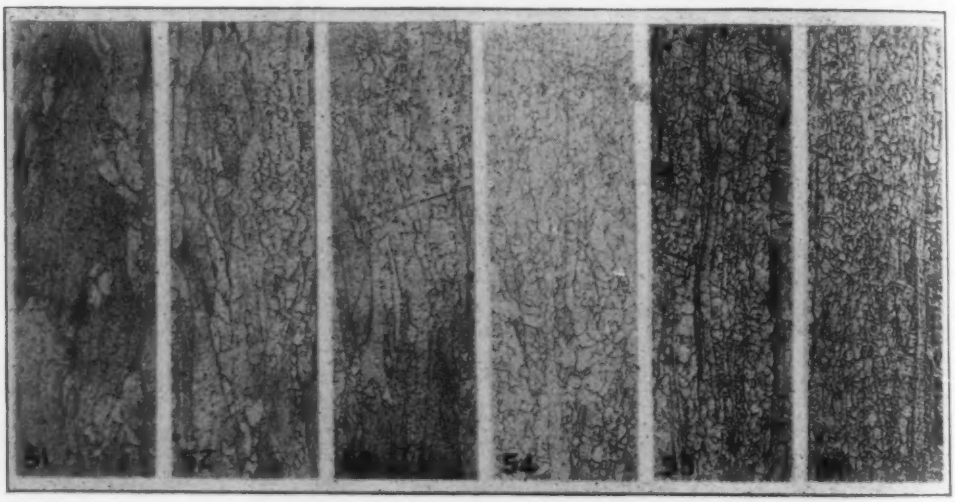


Fig. 15—Structure of Cold-rolled Electrolytic Iron Annealed ½ Hour at 500 Degrees Cent., 100 x. The Material of Fig. 10 Was Used.

ined at a magnification of 500 diameters, even in the specimens which had received the most severe reduction in cold rolling. The change in tensile properties noted between 750 and 930 degrees Fahr. (400 and 500 degrees Cent.) probably occurred between 840 and 930 degrees Fahr. (450 and 500 degrees Cent.). A microscopic examination was made of each of the tensile specimens from rolling series No. 3. It confirmed entirely the results shown by specimens from rolling series No. 1.

McAdam[2] has reported that ingot iron containing 0.034 per cent carbon, 0.01 per cent manganese, 0.008 per cent phosphorus, and 0.048 per cent sulphur (copper and oxygen not stated), required a reduction of 30 per cent before recrystallization was evident when annealed at 900 degrees Fahr. (480 degrees Cent.) for 8½ hours. A 40 per cent reduction followed by annealing for 16 hours at 950 degrees Fahr. (510 degrees Cent.) was required

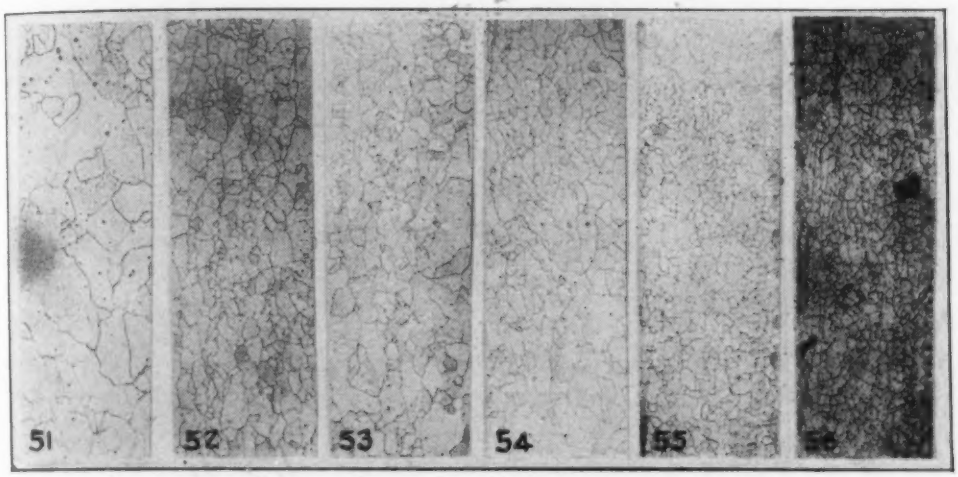


Fig. 16—Structure of Cold-rolled Electrolytic Iron Annealed ½ Hour at 600 Degrees Cent., 100 x. The Material of Fig. 10 Was Used.

in order to produce complete recrystallization. The temperature reported by McAdam is within the range, 840 and 930 degrees Fahr. (450 and 500 degrees Cent.), found for electrolytic iron, but in the present work recrystallization was noted in the specimen which had received a reduction of 44 per cent, which was held at temperature for only ½ hour. However, recrystallization of the electrolytic iron was not complete when annealed at 930 degrees Fahr. (500 degrees Cent.) for ½ hour, even in a specimen which had received a 90 per cent reduction by cold work. Had the specimen been heated for a longer time the recrystallization would undoubtedly have been more complete.

Chappell[3] reported incipient recrystallization at 350 degrees Cent. in a wire containing 0.07 per cent carbon which had received 89.2 per cent reduction in area by cold drawing. The time at temperature was not stated. The general appearance of new

[2]McAdam, D. J. Annealing Temperature and Grain Growth. *Proceedings*, American Society for Testing Materials, Vol. 17, pt. 2, p. 59, 1917, and Vol. 18, pt. 2, p. 69 (1918).

[3]Chappell, C. The Recrystallization of Deformed Iron. *Journal*, Iron and Steel Institute, 1914, Vol. 89, pt. 1, p. 460.

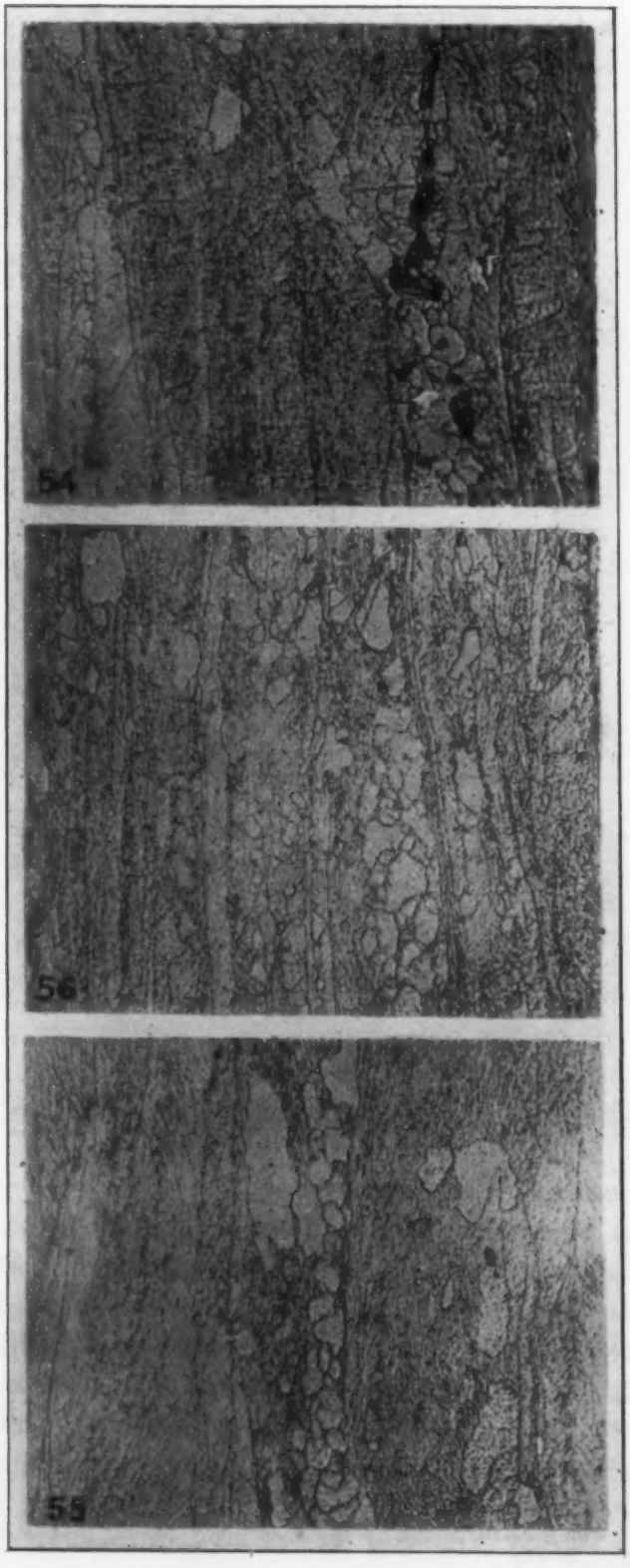


Fig. 17—Microstructure of Specimens of Cold-rolled Electrolytic Iron Annealed for 24 Hours at 360 Degrees Cent. and Then for 24 Hours at 425 Degrees Cent., 500 x. Etching Reagent: 2 Per Cent Nitric Acid in Alcohol. Specimen 54 Was Etched in Boiling Potassium Permanganate Solution for 30 Minutes, After HNO_3 Etching. The Darkened Particles are Cementite.

crystals, however, was not apparent until the severely worked wire was annealed at 930 degrees Fahr. (500 degrees Cent.), at which temperature, he stated, "normally shaped allotriomorphic crystals were easily visible at high magnifications." It is well known that increase in the heating period at the annealing temperature lowers the apparent temperature of recrystallization of a cold-worked metal, as determined by the first appearance of new grains visible with the aid of a microscope. In view of the relatively low temperature of incipient recrystallization reported by Chappell, specimens of electrolytic iron (Lot 5) obtained from samples from rolling series No. 1 were annealed for 48 hours between 660 and 680 degrees Fahr. (350 and 360 degrees Cent.). Microscopic examination of these specimens under a magnification of 500 diameters did not reveal the presence of any new grains. The specimens were then placed in the salt bath at 800 degrees Fahr. (425 degrees Cent.) and maintained at this temperature for 24 hours. Under a magnification of 500 diameters marked recrystallization was apparent in specimen Nos. 54, 55, and 56, as shown by Fig. 17. A few minute and poorly defined grains were to be seen in specimen No. 53 which had received 70 per cent reduction in thickness (Table II) but no distinct grains were visible in specimen Nos. 51 and 52, which had received the least amount of cold work. No. 52 had received a 63 per cent reduction. The conclusion may be drawn that a reduction of approximately 70 per cent is necessary in order to cause incipient recrystallization in electrolytic iron in the form of thin strip at 800 degrees Fahr. (425 degrees Cent.).

Similar specimens from Lot No. 5, rolling series No. 1, were annealed at 750 degrees Fahr. (400 degrees Cent.) for 24 hours and examined for evidence of recrystallization. At a magnification of 500 diameters no recrystallization was apparent in any of them. Specimen No. 56, which had received a reduction of 90 per cent, had at this magnification a slightly different appearance in certain areas. Under a magnification of 1000 diameters these areas were found to contain very minute but poorly defined grains (Fig. 18). The results were not definite enough, however, to warrant the conclusion that recrystallization had begun.

There were visible in all specimens, small grains of cementite which were often very similar to, and difficult to differentiate from, new ferrite grains forming at the crystal boundaries of the

severely worked metal, as may be seen in micrograph of specimen No. 16, in Fig. 11. Where there was any doubt as to the nature of the grains, the specimen was etched in potassium permanganate solution after the etching in nitric acid, whereby the cementite particles were darkened in contrast to the ferrite grains.[4] In Fig. 17, the micrograph of specimen 54 illustrates the results of

Fig. 18—Electrolytic Iron Reduced in Thickness 90 Per Cent by Cold Rolling and Annealed at 400 Degrees Cent. for 24 Hours, 1000 x. Etching Reagent: 2 Per Cent Nitric Acid in Alcohol.

this double etching. The light areas are the newly formed ferrite grains and the black areas the cementite.

IV. Discussion of Results

The results of the tensile tests and the study of the microstructural changes with increase in annealing temperature show that electrolytic iron of the composition used underwent marked recrystallization between 840 and 930 degrees Fahr. (450 and 500 degrees Cent.), whereas the open-hearth steel did not show marked recrystallization until annealed at a temperature somewhere between 930 and 1110 degrees Fahr. (500 and 600 degrees Cent.). This difference in the respective recrystallization temperatures is probably due to the relative purity of the electrolytic iron. It has been established by other investigators that impurities raise the temperature of recrystallization of a metal.

The presence of hydrogen in the electrolytic iron naturally

[4]Groesbeck, E. C. Metallographic Etching Reagents III; For Alloy Steels, Bureau of Standards Scientific Paper 518. 1926.

suggests itself as a possible factor in accounting for the difference in behaviour of the two materials. Pilling[5] has shown that hydrogen begins to be evolved at 195 degrees Fahr. (90 degrees Cent.), and that the maximum rate of evolution is at 390 degrees Fahr. (200 degrees Cent.), although the action is not fully complete until the iron has been heated above the A_3 point. Reed[6] reported that electrolytic iron (cathodes) became malleable at 650 degrees Fahr. (343 degrees Cent.) and recrystallized at 1245 degrees Fahr. (675 degrees Cent.). Rawdon[7] found, in a study of the effect of hydrogen on the thermal expansion and thermal analysis of iron, that any irregularities in the heating curves below A_2 disappeared after the first heating.

The electrolytic iron used in the experiments reported here was deposited in tube form and then annealed sufficiently to make it malleable. The strip prepared from this tube in the manner previously stated, after a light reduction by cold rolling, was given an anneal at 1400 degrees Fahr. (760 degrees Cent.) previous to the cold rolling tests referred to earlier in this report. Therefore, it does not appear possible that sufficient hydrogen could be present in the material used in these tests to have influenced the properties of the strip.

The tensile tests showed a gradual decrease in tensile strength with increase in annealing temperature below the temperature at which pronounced recrystallization occurred. This decrease in strength was probably associated with the first internal changes in structure occurring in the metal resulting in the birth of new grains sub-microscopic in size. At the lower annealing temperatures there was also probably a release of the internal strains which are set up in a metal by cold working.

V. Summary and Conclusions

A study was made of the respective temperatures of recrystallization of electrolytic iron and open-hearth steel especially selected for deep drawing operations. The study is a continuation of a previous investigation of the comparative cold rolling characteristics of the two types of material in which the electrolytic iron

[5]Pilling, Norman B. Effect of Heat Treatment on the Hardness and Microstructure of Electrodeposited Iron. *Transactions*, American Electrochemical Society 42, p. 9, 1922

[6]Reed, E. L. An Investigation of the Properties of the Iron Carbon Alloys, *Carnegie Scholarship Memoirs*, Iron and Steel Institute, Vol. 14, 1925.

[7]Rawdon, Hidnert and Tucker. Some Effects of Hydrogen on Iron. Forthcoming Paper.

was found to be slightly superior. Tensile tests and microstructural studies were made of strip given different amounts of work by cold rolling and annealed at progressively increasing temperatures.

The results confirmed the fact which has been well established by previous investigators that a marked decrease in tensile strength of cold-rolled metal occurs upon annealing which is accompanied by as marked an increase in elongation.

The noticeable difference in the recrystallization temperatures of the electrolytic iron and the open-hearth steel, has not been discussed in the literature, so far as the author knows. It was found that electrolytic iron underwent a marked recrystallization when annealed for ½ hour at 930 degrees Fahr. (500 degrees Cent.), as revealed by the microstructure and by the change in tensile properties. Incipient recrystallization was also noted in a specimen annealed for 24 hours at 750 degrees Fahr. (400 degrees Cent.). In the open-hearth steel marked recrystallization was observed in specimens annealed at 1110 degrees Fahr. (600 degrees Cent.), while no marked change was observed in specimens annealed for ½ hour at 930 degrees Fahr. (500 degrees Cent.).

The conclusion may be drawn that electrolytic iron of the composition studied has a lower temperature of recrystallization than special selected open-hearth strip steel. While the electrolytic iron used was fairly low in impurities, it is by no means pure iron. The recrystallization temperature of really pure iron would probably be different from that found for the electrolytic iron used.

The author would express his appreciation of the assistance of R. D. France, who made all of the tensile tests, and of J. W. Brewer, who prepared all the specimens for microscopic examination and made the micrographs.

STRESS-STRAIN CURVES AND PHYSICAL PROPERTIES OF METAL WITH PARTICULAR REFERENCE TO HARDNESS

By H. P. Hollnagel

Abstract

The physical properties of metals as associated with the stress-strain curve, are discussed and interpreted in terms of atomic forces of restitution, slip of an atomic nature and in mass. Interfering material is set forth as being the amorphous phase and hardness is defined as an average property of all the physical properties which are expressible in terms of the stress-strain curve, a means being suggested of correlating various hardness numbers. A line of research has been suggested to the end that a proper solution of this problem may be obtained. It is furthermore pointed out that hardness from the viewpoint of an average property may be defined as the work per unit volume per unit strain, and that this definition has the physical dimensions of pressure.

The Stress-Strain Curve

SUBSTANTIALLY all commercial metals are composed of coherent crystal fragments oriented more or less at random. These fragments vary in size from easily visible dimensions down to sub-microscopic magnitudes. The common boundaries of adjacent grains are regarded as possessing characteristics of amorphous or imperfectly crystallized material. Such conglomerates will, under a stress applied in any one direction, manifest deforming tendencies which accordingly are different in character, depending upon the aggregates affected. (See Fig. 2). If AB be the direction of applied stress, it is resolvable into two components, one AC along atomic planes, the other BC perpendicular to them. The one causes slip along these planes, the other rotation about centers within the individual crystals.[1]

As a result, amorphous material at the boundaries is either in tension or compression. Because of this and the irregular spac-

[1]The diagram herein given is in terms of two dimensional considerations, though the resolution of forces may be made equally well in three dimensions.

A paper presented before the Hardness Testing Symposium of the Society at the Cleveland Convention, September, 1925. The author, Dr. H. P. Hollangel, is physicist, Thomson Research Laboratory, General Electric Company, Lynn, Mass.

ing of the atoms, atomic forces of restitution are either those of attraction or repulsion. In either case the resulting action is that of restoring moments acting on individual crystals to cause their return to original positions. As atomic slip occurs at lower stress values than does slip in mass, and since the distribution of stress in any metal test piece is non-uniform (thereby permitting of highly localized stress conditions), these conditions are operative in every stress consideration. The ever-present forces of restitution exercise the preponderating effect in the unbalanced condition of forces for low values of stress, but when the stress is of such a magnitude as to exceed the equilibrated values of the restoring forces it definitely overcomes them, and slip occurs. Consequently, the restoring couple, together with atomic or block slip, gives a satisfactory means of formulating physical concepts of various properties in terms of a stress-strain diagram.

In no case does any portion of a stress-strain curve of an ordinary solid, like a metal, represent a particular physical property alone. In order to make this clear we will discuss (a) elastic, (b) hysteretic, and (c) permanent deformation. (Fig. 1).

The first of these causes a displacement of material in conformity with Hooke's Law and is viewed as being the result of an attempted alignment of the different aggregates along the lines of stress. (Fig. 2). This results in an interference at intercrystalline planes, causing interpenetration if in compression, or dilation if in tension of atoms during stress (whether of chaotic or regular distribution), and elastic return consequent upon removal of that stress. The effect takes place in a direction at 90 degrees to that parallel to the atomic planes of the crystals; it exercises the major influence outweighing such other conditions as may exist in the early stress history of a material; for example, it is known that slip occurs along atomic planes to some degree simultaneously with elastic deformation; it is, however, insufficient to completely change the slope of that portion of the stress-strain curve which is due to elastic deformation.[3] Such slip[4] occurs below the proportionality limit, but there is not a sufficient volume of material thus affected because of the interfering forces of neighboring aggregates which act hydrostatically to prevent such

[3]H. H. Lester, Ph. D., has very kindly placed at the writer's disposal information confirming this view.

[4]For a discussion of slip interference theory of hardening metals see Jeffries and Archer, "Science of Metals," McGraw-Hill, 1925.

slip except at positions where the material is not contained.

Second, as the crystalline masses become more coerced from the positions corresponding to their respective orientations, slip takes place to a greater degree, with gradual decrease in the interfering forces consequent on, the stress not only equaling, but locally exceeding the atomic forces of restitution. Moreover, the elastic interfering forces are no less locally than before, but owing to the dispersion of their effort and the readjustment of atomic

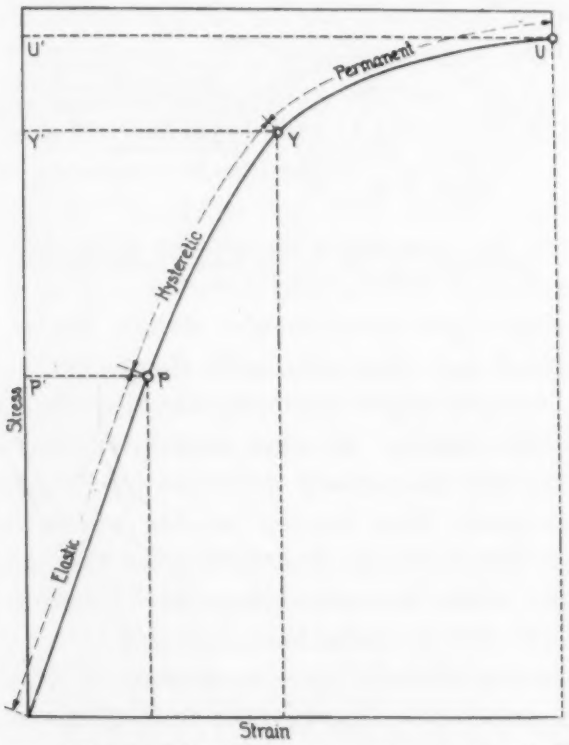


Fig. 2—Schematic Representation of Chaotically of Deformation.

planes to new positions of partial equilibrium, on the average their result is less and an increase of slip along these planes sets the material in part in new configurations. Directly above the elastic limit, therefore, there exists a transition region in which material after being stressed if permitted to rest tends to relieve itself of strain and return to its original state of perfect elasticity. As a matter of fact, it never does completely recover.

It has been pointed out that as the stress comes on, there is a gradual tendency toward alignment; there also results a slip effecting a departure from perfect elasticity; simultaneously the

interfering forces tend to prevent the rotation of crystal blocks. If, therefore, the components of tensile stress which act transversely to atomic planes of crystalline aggregates overcome the repellent and attracting forces of the atoms, there is only one result as the stress increases, i. e., more slip. This is equivalent to the statement that the longitudinal stress along atomic planes becomes greater as the resultant transverse stress diminishes. The greater the penetration into or the dilation of the atoms from one another, the more do forces of atomic repulsion and

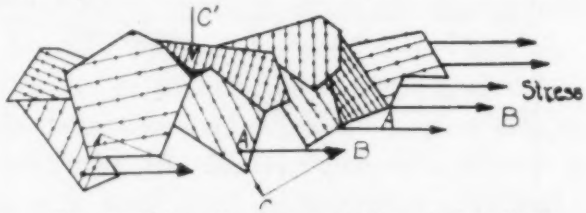


Fig. 2—Schematic Representation of Chaotically Distributed Crystals and the Forces Tending to Cause Slip and Rotation of Them.

attraction exert themselves within certain limits; consequently once equilibrium has been attained, there will be greater grain interference, i. e., at right angles to the direction of further slip along the atomic planes. In this manner strains resulting from interference become permanent as the latticed aggregates are distorted or destroyed, they having become welded one to another through their distortion or destruction in the amorphous phase. This represents what has taken place at C′ under permanent deformation when slip in mass has occurred.

Before passing directly to a consideration of the stress-strain curve and its relation to hardness a few of the physical properties which relate to materials will be examined.

Physical Properties

The three points characterizing a stress-strain curve are: (1) the proportionality or elastic limit—P; (2) the yield point—Y; and (3) the ultimate limit—U. In the light of what has been said, they have the following significance: P, (Fig. 1), represents a point in the strain history of a specimen of metal just beyond which slip on atomic planes outweighs the transverse forces of restitution previously referred to. In a perfectly elastic material, no such point exists, its restoring forces are always 100 per cent effective, whereas, for an imperfectly elastic substance, immediately above this point, a sufficient volume of material changes

its configuration, in the aggregate, causing the average restoring moments of the forces of restitution to be insufficient for immediate recovery to the original state. Below this point the material returns approximately on removal of the stress.

The point Y indicates that slip on atomic planes and in mass has so far outweighed the forces of restitution that the configuration of separate aggregates forming a certain volume of the test piece has been progressively and definitely distorted or destroyed. The major forces of restitution of the individual aggregates have, therefore, become ineffective and another crystal arrangement has resulted. Below this point, considered from a macroscopic point of view, the material will, in time, resume its original dimensions to a very considerable degree.

The point U is the stage at which restitution is zero and atomic slip and in mass has proceeded to such an extent that the so-called keying action between slip planes has caused complete destruction of same. Such a volume of the material has been destroyed that shearing is effected at the crystal surfaces with the result that rupture occurs. Not only have crystal blocks been actually ruptured, but fracture has occurred at crystal boundaries. This may be seen in Fig. 3[5] by close examination of the line of fracture.

Brittleness. When P, Y and U are of such values as to be practically indistinguishable a substance is said to be brittle. In the words of the tester, when elongation per unit length beyond the proportionality limit is exceedingly small, the departure from Hooke's Law is negligible and the material ruptures at the same stress as that at which it yields. See Fig. 4, Curves IV and VII. This requires that the atoms or aggregates to be so firmly packed and interlocked that a condition of destruction is imminent under the slightest displacement from the physical conditions existing at P. A locally applied stress of considerable magnitude may cause the internal strains necessary in such cases. A condition of highly strained comminuted crystals with considerable amorphous material included, as previously described, would account for this condition in a metal. Glass is an example of a brittle amorphous substance.

Ductility. When Y is removed from P, and U is markedly distant from Y, then the material is ductile, for though elastic after-

[5]Furnished through the kindness of Dr. H. H. Lester.

effects cause a gradual readjustment of the test piece, this is so slight compared to the change which permits of permanent distortion of the specimen, that it is insignificant in the final result. (See Fig. 4, Curves V and VI.)

In a material like copper, P and Y are quite close together, while Y and U are relatively far apart. The strain per unit stress applied is much greater than is the case just below

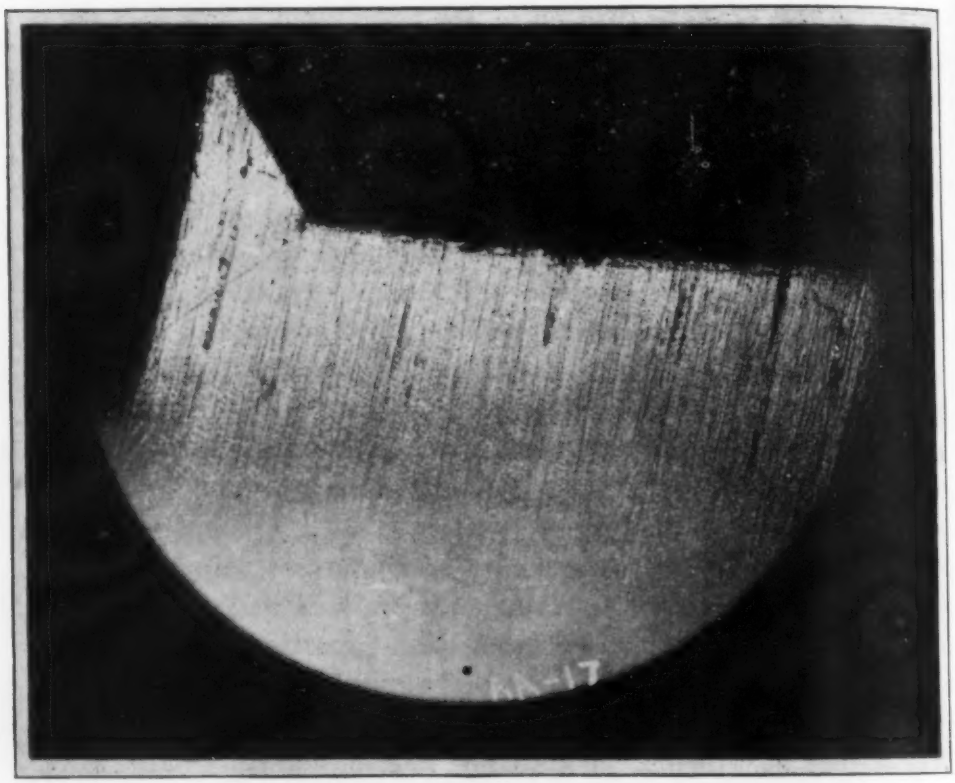


Fig. 3—Photograph of a Bar of Steel Showing the Manner in which Fracture has Occurred at Crystal Boundaries.

the proportionality limit. In ordinary materials, this necessitates that the substance slip along atomic planes to a greater extent than in non-ductile materials. This also requires that in the process of stressing a specimen, the latticed aggregates be larger than those in which there is less slip. In other words, the disruption of material, with the formation of smaller crystals does not take place until these planes have been sufficiently destroyed by keying action between them to form the interfering material which, in a large measure, causes restitution and in this case simultaneously toughens the same. If the stress is interrupted, restoring forces

cause a redistribution of the material, which modifies its physical characteristics. As a result, Y and U may not be far removed from each other, but are elevated in value. While paradoxical it therefore appears that ductile materials become more resistant to stress by reason of their ductility, i. e., the destruction of larger aggregates in the stressing of materials causes the small crystals

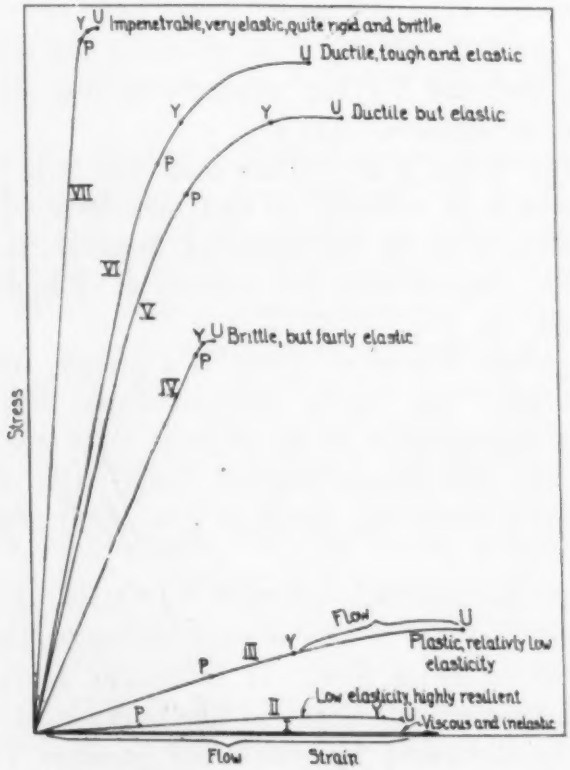


Fig. 4—Stress-Strain Curves Indicating Relative Physical Properties.

to form; these in turn interfere more and more, causing slip in mass as here explained. If the lattice be perfect and under stress and the proportionality limit be exceeded, a condition of chaos results within the crystalline structure and the ductility of the material becomes less as a consequence.[6]

Resilience. A body which, under stress, has its dimensions considerably altered, but on release of the stress resumes its original dimensions immediately without trace of permanent set, or very approximately so, is said to be resilient. This means that in the stress-strain diagram (Fig. 1, the distance between

[6] W. S. Davey has shown this to be true for a perfect lattice of copper.

P and Y may be relatively large and that between Y and U small. It also requires that much energy be stored in the distortion of the body. Viewed from the angle previously presented, such material as is present in the randomly oriented lattice form, may slip along the direction of atomic planes to new positions of equilibrium (below Y), but is insufficient in volume compared to that which acts to interfere with the aligning action of the stressing forces, consequently when the latter forces have reached sufficient great values, a marked volume of the material not only slips suddenly in mass, (beyond Y) but ruptures almost simultaneously at inter-crystalline surfaces. (at U).

While rubber is not a crystalline material, it is indicative of the characteristics here referred to and manifests resilience to a high degree. Because of its low modulus, however, the strain per unit stress is much higher than for a material of higher modulus. See Fig. 4, Curve II.

Plasticity versus Viscosity. Lead is a plastic, while pitch is a viscous material; one shows displacements proportional to the load applied and returns to its original state when the stress is removed within the proportionality limit; the other is displaced continuously under the action of a constant stress irrespective of its value.

Plasticity is characteristic of solid materials while viscosity is typical of fluids.[7] Pitch is harder than putty requiring greater compressive force to cause flow. It is brittle, moreover, when acted upon by impact stress. Glass, in like manner, is harder than lead, but may be deformed by continued pressure though it is fragile and brittle to locally applied or instantaneous stresses. The modulus of elasticity of glass is much higher than that of lead, but under the application of a continually acting stress, gradually applied, it will flow slowly. Continually increasing stress results in rupture without marked elongation.

Further difference between such materials seems to reside in the effects produced by stresses applied for different periods of time. A plastic material may have the stress active for an indefinite period and produce no appreciable permanent deformation within the elastic limit, whereas a viscous substance, no matter how hard it apparently may be, will show greater deformation

[7]Maxwell's Theory of Heat, page 296.

the longer the stress is applied, having no elastic limit.[8] Substances like glass show viscous characteristics over long periods of time, though they may be brittle to ordinarily applied and impulse stresses.

If any plastic or viscous material be contained on all sides, and pressure applied, it will act like an incompressible fluid.[9] In fact, Dr. Bridgeman, of the Jefferson Physical Laboratory at Harvard University, has caused glass to flow under high pressure, showing it to have the viscous characteristics of a fluid. A plastic material is represented in Fig. 4, Curve III, a viscous material in Fig. 4, Curve I.

The question arises—how can a viscous material also appear to be brittle? If the period of application of the stress is of the nature of an impulse time, and is of such an order of magnitude that there results little or no inter-penetration or dilation of the atoms which make up the closely packed structure of the substance, then there will be no apparent give to the material. To all appearances, the body will act instantaneously as perfectly rigid material. The forces of resistance which characterize inter-penetration have magnitudes due either to external or internal stress such that the material while locally ruptured by impact still retains its contour and appears to be rigid. In fact atoms penetrate others until equilibrium has been attained and beyond this stage their configuration is actually destroyed by the shearing action of the stress.[10] When this occurs in quantity rupture has occurred locally.

Finally, it requires time to produce the gradual effects known as flow, which in plastic materials, occur as the result of slip between crystalline planes; such substances, when impacted, will act brittle and locally rupture, unless so plastic as to flow under that kind of stress in the time of impact, in which case, they possess relatively low moduli of elasticity (lead). Materials that are irregularly close-packed, do not have the capacity for absorbing shock readily in this manner, therefore, under impact they must have (1) either sufficient strength to withstand a bending stress locally,[11]

[8] Local citation.

[9] Bridgeman, *Proceedings* of the Jefferson Physical Laboratory.

[10] This effect brings cogently to mind the action of a strong non-uniform magnetic field on a sheet of metal which is being forcibly removed from or inserted into it; as is well known, because of the induction, there results a mechanical reaction tending to prevent either. It is almost impossible to accomplish the removal or insertion at high speeds without the expenditure of large quantities of energy.

[11] In rebound devices used for testing hardness the materials tested are hydrostatically contained at the point of stress and, therefore, withstand local stress well.

in which case they are not affected by impact; (2) not having the capacity to glide or flow under the action of impulse forces, or (3) being wholly amorphous they fracture. In a crystalline material slipping would occur on interatomic planes, or intercrystalline surfaces under such stresses, particularly if the crystalline blocks which constitute the structure are hydrostatically contained and the forces localized. Ordinarily the last condition is not true in most large impact testing devices and as it takes time to displace even the smallest mass, the phenomenon of flow does not take place to any marked degree. This being the case a reasonable explanation is offered why atoms take the shock rather than blocks of crystals as manifested by their displacement in mass.

To be sure, if the magnitude of such stresses be sufficient, the time required may be less since large enough forces then cause impact effects of deformation and materials that are contained may be caused to flow at the surface of application of the stress.

When neither slip nor actual rupture take place, to an appreciable extent in a metal, the reaction evidences itself as a reflected wave, which is dissipated as heat within the material. Under such conditions the displacements are all elastic and the wave will be an average reaction measured by the energy returned when stress is released.[12] This being the case the phenomena of flow, taking place for gradually applied loads, has no appreciable effect and the strain resulting per unit applied stress is markedly lower than is required at the elastic limit. The material, therefore, appears to act as though it had a higher modulus because of this lower value of the strain resulting as a consequence of the short impulse time. (Fig. 6.)

Therefore to impact stresses most materials act more rigidly than to gradually applied and increasing stresses. Most of the more rigid metals are apparently the same in this respect. It would seem that all metals from those of low to those of high moduli of elasticity would show apparently higher moduli under impact.

Toughness. If a material has a high ultimate limit, relatively high modulus and a considerable ductility, it will be represented by a curve of the form shown in Fig. 4, Curve VI. If, with this ductility, the elastic limit of the resulting material has been raised, as has previously been indicated, it has a structure such that for

[12]The time of return of this energy will depend upon the velocity of sound in the material.

stress beyond the elastic limit, the interference with slippage along atomic planes is great. Slip in mass of the material occurs notwithstanding the interference offered and rupture occurs when the clashing, due to the keying effect, outweighs the interference of restitution despite its magnitude.

Impenetrability and Malleability. A material, to be impenetrable must have high values of the modulus of elasticity, proportionality limit, little or no plasticity, and a high ultimate limit. It must be rigid. A metal, to be penetrable, must be plastic; have low modulus, proportionality and ultimate limits. This means that it would also be malleable. A high modulus requires greater effect of the forces of restitution per unit applied stress, therefore, fine grain structure. Under these conditions, slip in mass and atomic planes, would be reduced to a minimum, which makes both yield point and ultimate high. The opposite will be true for malleability. (See Fig. 4, Curve VII.)

Elasticity. Most bodies in nature are perfectly elastic to some degree, i. e., to a certain extent return to their original physical state from that of strain when the stress has been released. The degree of their perfect elasticity is largely determined by the value of their modulus of elasticity. Furthermore, they are not particularly characterized by resilience, in which case the magnitude of the strain per unit stress applied is very large, as in rubber.

The most perfect elastic material would be one in which all atoms would be either in tension or compression and would have forces of restitution which were dependent only upon the complete interpenetration or dilation along atomic radii. Such material would always manifest 100 per cent coefficient of restitution for all strains; but substances in the industry are perfectly elastic only to the proportionality limit.

Is rubber elastic? It is, but to a very much less degree than a material like glass or quartz. The last mentioned are moreover brittle and show practically no resilience. Steel may be elastic, and ductile or brittle, depending on its heat treatment; in like manner a material like copper may be elastic up to a certain stress but plastic beyond that. The stress characterizing this point is the proportionality limit P. (Fig. 1). Therefore, an elastic material is one that has not only a high value of Young's modulus of elasticity, but also of the proportionality limit.

Whereas these materials all return to their original shape upon removal of stress within the proportionality limit, the displacement per unit stress in rubber is much the greater, i. e., it has relatively higher resilient qualities.

Beyond this point, every stress and strain corresponding thereto indicates an instantaneous value of the elasticity that characterizes a material momentarily. This is the result of a redistribution of the crystalline conglomerate caused by crystal rotation resulting in atomic and inter-crystalline slip and atomic penetration. Consequently, every point beyond P is indicative of another material caused by what in practice is ordinarily referred to as cold work. When sufficient of the material has actually yielded there will be relatively slight return to the original state on release of the stress, and as previously referred to the elastic after-effect will be of little consequence. Such metal, if allowed to rest for a time will, upon re-application of the load, exhibit values of P, Y and U, in excess of previous ones.

Modulus of Elasticity. The modulus of elasticity has been shown to be an important factor in all these discussions. There is no material either perfectly rigid or soft in the absolute sense, for all substances lie between two modular extremes (See Fig. 5); again, it has been shown that no medium acts directly in accordance with Hooke's law throughout its entire stress-strain history, but has other physical properties, such as yield point, proportionality limit and ultimate limit, which in turn determine the characteristics know as brittleness, ductility, (plasticity) toughness, resilience, rigidity, impenetrability, elasteity, malleability.

There is one quality that has been omitted—hardness; it has not been used because no simple way of presenting it has been divulged. On purely theoretical considerations, it should be associated with rigidity and therefore everything should be classed between materials having zero and infinite moduli of elasticity. This would be ideal were everything elastic and Hooke's law obeyed throughout the stress-strain history, but as this is not true rigidity alone as a criterion of hardness is impossible.

Rigidity. If a stress, whatever its magnitude, produced zero strain in a body, the modulus of that material would be infinitely great; it would be perfectly rigid, the hardest material, with an elastic limit of infinity, and neither yield point nor ultimate limit would be measurable, for they would be infinite in value. Most

materials in nature are elastic to a degree and are, therefore, deformed within the proportionality limit in conformity with Hooke's law. The case where a minute force produces a very large strain, is when the modulus of elasticity approaches zero as the force diminishes in magnitude to the other extreme. This would be indicative of the softest material. A material having this characteristic would be the one that had zero modulus, zero elastic limit, no rigidity, yield or ultimate limits.

The moduli of two such materials, neither of which exists, would be represented by the coordinate axes as shown in Fig. 5.

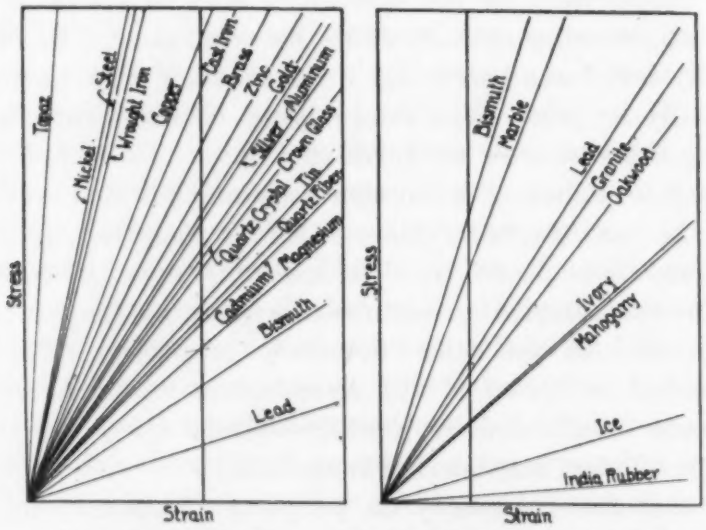
Most engineering substances lie between these. In representing the physical facts herein set forth any one may be characterized partially at least by a curve lying between these extremes and having the character set forth in Fig. 4. There are shown in Figs. 5 and 6 moduli of a number of materials and as has been indicated the most impenetrable and softest materials have infinite and zero moduli respectively. The modulus alone, however, is insufficient to characterize an engineering material from a technical standpoint, and as the other necessary properties have already been examined in terms of the stress-strain curve it seems that proper means is afforded by the stress-strain curve for the characterization of most engineering materials.

It is therefore necessary in properly characterizing such a material to define it in terms of effects, depending upon the chemical and physical i. e., the atomic and crystalline, microscopic and macroscopic constitution of the material.

Hardness as an Average Property. A little reflection will establish this view, particularly after considering (1) the elastic deformation due to the interpenetration of the atoms in crystals; (2) the deformation characterized by "elastic after-effects," and (3) that due to slip on atomic planes and in mass, which give the yield point and ultimate limit due to the "keying" of crystals. As pointed out these phenomena cannot be considered as taking place independently and in succession as the stress-strain curve would seem to indicate, but rather simultaneously. As one increases, its effect exceeds the others and the result observed, while an average, is determined by the major influence. An average property which includes the effects of all other characteristics included in the stress-strain curve is hardness. It is an average resistance of the material to deformation of any kind, for the

resistance, whether of restitution or keying, has different values throughout the specimen. This hardness may be divided into two kinds, that characterizing the surface and that due to cold work. In either case it represents a physical resistance, due to an average of the resultant forces of restitution of that volume of material which is immediately affected by the working stress.

It may be defined in terms of the stress-strain curve as this gives a record of the necessary physical facts for a particular piece of material. (Fig. 7).



Figs. 5 and 6—Curves Showing Moduli of Elasticity of Various Materials.

The area under this curve is indicative of the energy required to stress a certain volume of the substance to the ultimate limit. It is characterized by three parts, O B P and B P Y D and D Y U C. The first is the deformation energy necessary to stress the material at the surface to the elastic limit. The second, the energy which characterizes the work of deformation beyond the proportionality limit to the yield point and may be termed the energy of hysteresis. The third, the energy which characterizes cold work, i. e., between the yield point and the ultimate limit. If, therefore, these be summed up, there results the energy used in overcoming three effects. This acts on a certain volume of material developing a strain. If distributed uniformly over the entire strain a quantity results which is related to all its engineering properties. It therefore appears that solid substances have a particular property characterized by the work

done on them per unit volume per unit strain for the applied stress. This property has different values, depending entirely upon what portion of the stress-strain curve is examined. At P, for example, its value is

$$\frac{(OP)\ \ (OB)}{2(OB)} = \frac{OP}{2}$$

At Y it has the value given by the

$$\frac{\text{area } OYD,}{OD}$$

and so on. In this manner a value may be obtained which is characteristic of any portion of or the entire stress history of a material. This the writer has called the average hardness. It is a definite physical quantity having the dimensions of pressure

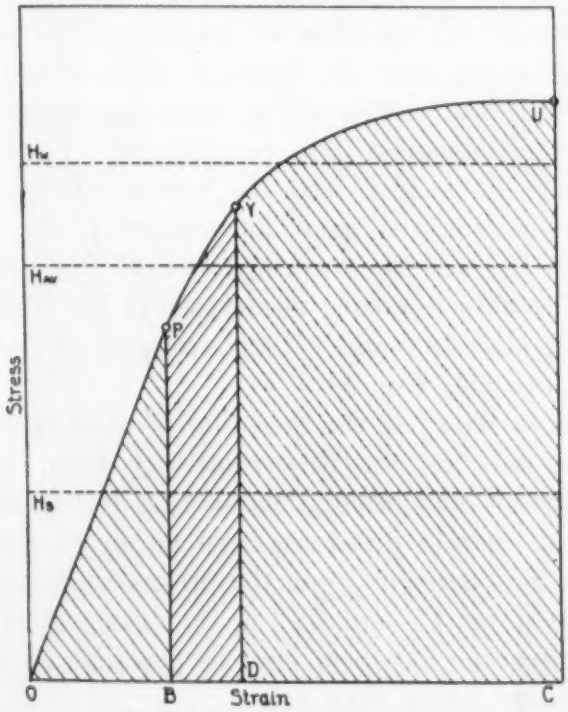


Fig. 7—Hardness Characterized by the Energies of Elastic and Permanent Deformations.

$[ML^{-1}\ T^{-2}]$ being the work necessary per unit volume per unit strain to overcome the average of the forces of resistance of all the atoms participating. This in the last analysis also characterizes the different hardness numbers referred to in practice. It is termed the average hardness of the material not only because

it is an average of the atomic forces of restitution with which we are concerned but also because the work of deformation is cumulative in effecting a resultant strain. This, when divided by that strain, must give an average value. It is equivalent to constructing a rectangle of such a height as to give the same area as that under the curve having the strain as its base.

As will be seen, it provides a possible means for comparing various types of hardness numbers. These are determined by devices which do work while elastically or permanently deforming material in a non-uniform manner, and as there is a stress-strain diagram characteristic of every solid of given dimensions, a comparison of their numbers resolves itself into a determination of the volume of stressed material and the magnitude of the strain. Of these, the former is the more difficult.

The area under a stress-strain curve varies in character with the magnitudes of (a) modulus of elasticity, (b) proportionality limit, (c) yield point and (d) ultimate limit of the material. It is understandable that when all of these are high a large value H_1 of the average hardness is to be expected. (Curve I, Fig. 8). If the value of (a) is high and the values of (b), (c) and (d) are such as to flatten the curve at relatively low values of the stress, a value H_2 will be lower (Curve II, Fig. 8) than when there is considerable ductility accompanied by a lower value of the modulus of elasticity and a high value of the proportionality limit—H_3 (Curve III, Fig. 8).

On the other hand, if the modulus of elasticity is low, as in certain metals, and the ductility relatively high, with comparatively low value of (b), then a value like H_4 or H_5 will be obtained which is entirely different from either of the other three. As referred to, this depends upon the ultimate limit and the value of the strain (Curves IV and V, Fig. 8) from which it is evident that values of average hardness may be obtained to fit almost any of the many cases which may arise.

All metals have approximate proportionality limits. This makes possible the determination of a quantity which is characterized by the modulus of elasticity and the proportionality limit of the material. This property would be determined by the work represented by the area of a triangle for each substance as divided by its base. For purposes of discussion, we may call this the "average surface hardness" of the material. It is characterized

by the depth of material associated with the surface which manifests its resistance to deformation by return to original dimensions when the stress is removed.

The ratio of the mutually perpendicular sizes of such triangles determines the respective moduli of elasticity of the various materials concerned. One-half of the stress ordinate at any proportionality limit P will be directly proportional to the surface

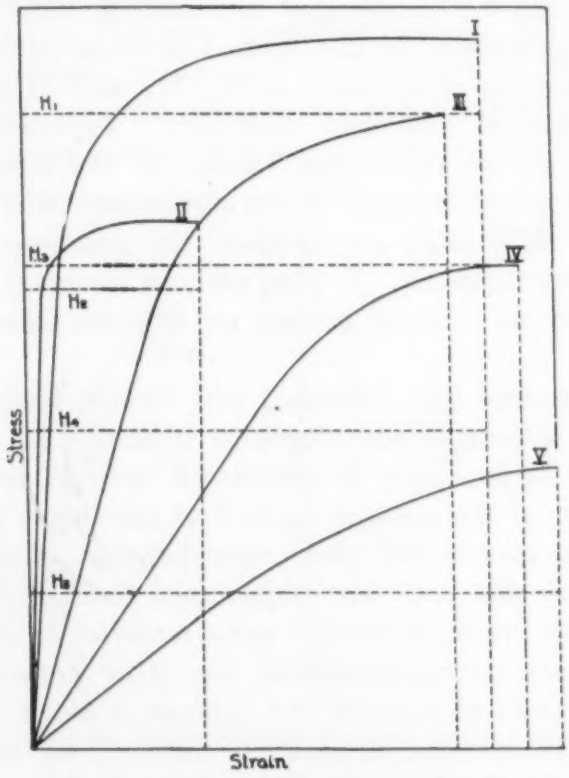


Fig. 8—Various Kinds of Stress-Strain Curves and the Average Hardness Characteristic of Them.

hardness of that material at that limit. Figs. 5 and 6 show a series of graphs of the moduli of several materials. In like manner the surface hardness of metals at any stress within the proportionality limit is directly proportional to one-half the applied stress. Therefore, for any given value of the strain (A) the average surface hardness numbers of metals should follow a relatively simple law. (Fig. 9).

The hardness as measured by any device in which the ultimate limit is arrived at for a portion of the material will be made up of two parts, one the average surface hardness here referred to; the other, that due to cold work. As shown in Fig. 7, the result-

ing average hardness as determined by indentation will therfore be made up of two factors, generally speaking, one characterized by the modulus of elasticity and the proportionality limit, the other by the proportionality limit, yield point, and ultimate limit which determine the cold work. As has been pointed out this requires a consideration of the same physical properties as are of importance in the determination of strength. The areas which characterize the work expended in reaching these values of the stress are represented by the shaded portions of the figure and the hardnesses are designated as H_s, H_w and H_{av}.

Correlation of Hardness Numbers. To correlate hardness numbers the stress, strain and volume of the stressed materials must be determined for each of the specimens tested by the different devices. As pointed out, of these, the measurement of volume will be the more difficult. It can, without doubt, be satisfactorily determined by an etching process on sections taken in different directions.

Let the indentation, rebound, and scratch methods be examined, and to eliminate one source of difficulty let it be decided that the size of specimen be considered such as to make impossible the effect of the anvil or contact of specimens with the anvil in such apparatus where their consideration is necessary. In scratch devices this is of less importance than in the other two. In general, the readings for the same material as determined by two devices, one operating within, the other beyond, the elastic limit will differ when reduced to a common scale; the latter being the higher. In rebound apparatus, however, an additional fact must be considered; because of the short time of application of the stress, the stressed material acts as though it has a higher elastic limit and modulus of elasticity and therefore, the resultant hardness number recorded will be higher than that which would be determined by the same stress applied at a lower time rate, as is the case in indentation apparatus (Fig. 10). Moreover if a rebound apparatus does not deform the material while an indentation device does, the relation of their hardness numbers should be proportional to the ratio of the

$$\frac{\text{area OBP}}{\text{area OYUCD}}$$

Fig. 7. But as the volumes affected are different in the two cases,

the stress per unit volumes must be determined in order to make feasible such a relation.

The scratch test is accomplished by causing an indentation by means of a tool which thereafter is drawn across the material. As this process progresses, the material ahead of the abrading tool is ruptured and falls away. Therefore, the reasoning herein contained applies in this case.

There is no concordant correlation between the several types of hardness numbers because of (1) the rate at which the work is done, (2) the amount of material under stress, (3) the magnitude of the stress and strain is so different in each of the methods

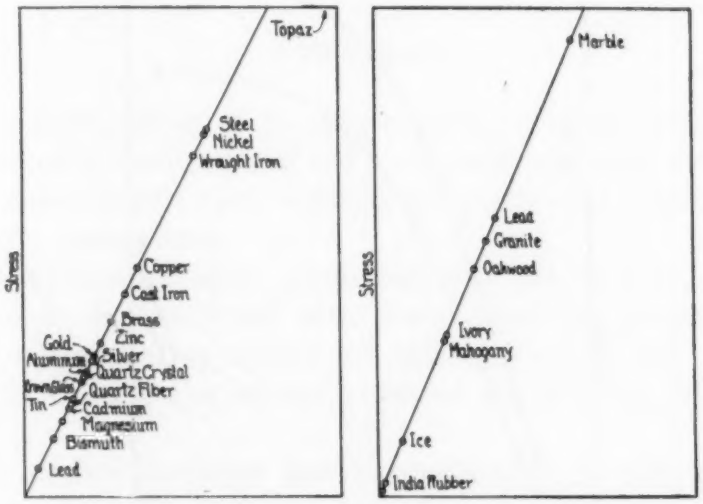


Fig. 9—Average Surface Hardness for Given Value of Strain.

utilized. In fact it would be quite extraordinary to expect a simple relation. In rebound apparatus where the energy is expended in either elastic or permanent deformation in a very short time (thousandths of seconds), the impulse forces are often excessive, being greater at times than the deforming stresses necessary for permanent set; but still the materials show no permanent deformation, because of the time required for such phenomena to establish themselves. A scratch, as the record left by a deforming process which permanently destroyed the surface of the material, may have been accomplished in a fraction of a second or several seconds; in either case the record is different, other things being equal. Indentation devices require time, usually a few seconds at least, to set the material permanently. So from the viewpoint

of time involved in the application of the stress there is a considerable degree of variability to be contended with.

In all instances the volume of material stressed is very different, and as the stresses are of entirely different magnitudes, the volume-stress intensity varies greatly in different directions in the several cases. Notwithstanding the apparent insurmountable difficulties, the stress-strain curve gives the necessary method to ascertain the relationship between the various hardnesses.

The variation of the modulus of elasticity, if any, proportionality, yield and ultimate limits with the duration of impact

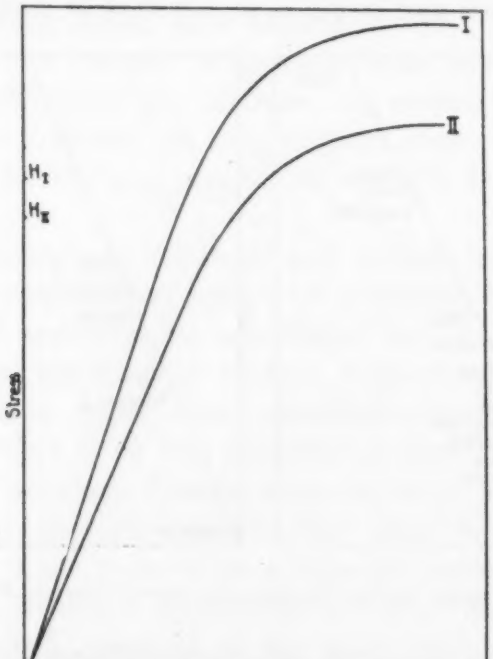


Fig. 10—Average Hardness Indentation Versus Impact.

times should be studied for different instruments, in which case the information leading to the possible solution of this difficult problem may be ascertained. Similarly the results of these values depend on the size of the specimen, and an investigation should be made from this angle.

It is not the intent or purpose of this work to establish a new hardness number or to attempt to cause a discontinuance of present usage. It is necessary, however, to establish a system whereby it will be possible to have the various numbers correlated

and interpreted in accord with a common terminology. It is therefore suggested that a definite system of units such as the Absolute (C. G. S.) or English (F. P. S.) system be adhered to.

If the "average hardness" be stated as a pressure expressed in pounds per square inch or in dynes per square centimeter, in any consideration, it must be maintained so throughout. In this manner and by division of all such hardness figures by a common factor, numbers may be obtained which are of a reasonable order of magnitude and that may be dealt with conveniently. Once these have been secured the problem resolves itself into a simple calibration for the several types of instruments used for measuring these numbers.

Conclusion

An interpretation of the stress-strain curve is given in terms of the forces of restitution of the atoms, slip—of an atomic nature, and in mass. Restitution depends on interfering material which is probably amorphous.

The various physical properties are discussed in terms of these factors as associated with the stress-strain curve.

An energy interpretation of this curve and the work performed in causing the several kinds of deformation has been set forth.

The surface hardness and hardness of cold work have been defined as average phenomena occurring during stress. The hardness of any material as defined by the stress-strain curve is an average phenomenon including the above hardnesses and requires a consideration of all other physical properties herein referred to.

This determination of hardness, from an energy point of view, includes its stress-strain history to that point on the stress-strain curve at which any device may be viewed as operating.

The definition of hardness has been given as the energy per unit volume per unit strain. This is a definite expression of the physical magnitudes involved and gives to hardness the dimensions of force per unit area $[ML^{-1}\ T^{-2}]$.

The definition of hardness given makes possible theoretically the unification of all hardness measurements; the difficult factors formerly met with may be satisfactorily explained by this concept of hardness.

A means for correlating the various hardness numbers has been formulated and work along this line is suggested.

The writer is much indebted and wishes to make grateful acknowledgment to his kind friend and colleague, Dr. H. H. Lester, for having so generously provided him with technical data and photographs; to E. A. Harty, of the laboratory, for having prepared the drawings; to the editorial staff of the *Iron Age* in whose publication[13] a portion of this work has appeared and to the editorial staff of the A. S. S. T., through whose kind co-operation this paper has been made possible.

[13] *Iron Age*, November, 1924; March, 1925.

Educational Section

These Articles Have Been Selected Primarily For Their Educational And Informational Character As Distinguished From Reports Of Investigations And Research

THE CONSTITUTION OF STEEL AND CAST IRON—PART II

By F. T. Sisco

Abstract

This article discusses the determination of critical points in metals and alloys and the method of plotting the cooling curves, showing the points where these constitutional changes take place. From the cooling curves of a large number of alloys of the same series the equilibrium or constitutional diagram is constructed. This diagram shows graphically all of the changes taking place in the alloys of the whole series. As typical of a simple series the cadmium-bismuth alloy diagram is constructed.

ONE of the distinguishing characteristics of a pure metal is that, almost without exception, it crystallizes when it solidifies from its melt. In addition it solidifies at a constant and definite temperature. If we heat pure tin in a crucible to about 600 degrees Fahr. (300 degrees Cent.) and let it cool slowly the temperature drop will be uniform until 449.4 degrees Fahr. (231.9 degrees Cent.) is reached. At this point the temperature of the metal remains constant until the whole mass is solid, even though the rate of cooling for the furnace is the same as before. Enough heat is liberated in solidification so that it practically balances the normal cooling of the furnace and crucible.

Conversely, if we heat tin in a crucible the temperature rises uniformly until 449.4 degrees Fahr. (231.9 degrees Cent.) is reached. At this point the temperature will remain constant until the whole mass is molten, even though heat is being constantly and uniformly applied. The heat absorbed by the metal in melting balances the heat being applied, so that there is no further

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increase in the temperature of the metal until all has melted. This energy used in melting is known as the heat of fusion. If this amount of heat is measured accurately it will be found that the same amount that was absorbed in melting is evolved in freezing (solidification).

This latent heat held by the atoms of the molten metal is released when they arrange themselves in definite geometrical patterns or crystals in freezing, and gives measurable evidence of its existence by a jog in the cooling curve.

We have noted in the previous article that when an element such as iron undergoes an allotropic change in the solid state, an evolution of heat in cooling or absorption of heat in heating occurs. This measurable display of heat is due to some change in the metal;[16] in some cases, to a change of position or grouping of atoms in the space lattice.

By correlating the heating and cooling curves which show the freezing (melting) temperature and the heating and cooling curves showing the temperature of the constitutional changes in the solid state, for a pure metal and for alloys of this pure metal with one or more other metals, we are able to construct an equilibrium diagram which is used as a basis for the intelligent study of the structure of metals and alloys.

Determinations of Constitutional Changes in Metals and Alloys

Much of our advance in the study of the constitution of metals and alloys has been due to the perfection of precision apparatus for the exact location of the changes taking place when they are heated or cooled. Equilibrium diagrams are a plotting of these changes for a whole series of alloys. The iron-carbon diagram is constructed from the results of a large number of cooling curves made on the alloys of iron containing from 0.01 to 6.00 per cent or more carbon.

The apparatus necessary for the determination of the freezing (melting) point of a metal or alloy is relatively simple. It is only necessary to have an accurately standardized thermocouple and instrument for reading the temperature. The evolution of heat at the freezing point is so marked that it leaves no doubt as

[16]We can properly term these internal changes which are evidenced by an evolution or absorption of heat as constitutional changes. This is better than allotropic changes, in view of the fact that the exact meaning of allotropy is doubtful.

to its position. It thus becomes only necessary to eliminate errors in technic or equipment. This is relatively simple.

In the determination of the points where allotropic or other changes take place in the solid state, where internal atomic rearrangement occurs, more refinement is necessary. Oftimes the evolution of heat in cooling or absorption of heat in heating is very faint, making it necessary to use delicate methods to detect the change.

In general the apparatus for the determination of constitutional changes in metals or alloys consists of an electrically heated furnace capable of uniform heating and cooling, one or more accurately standardized platinum, platinum-rhodium thermocouples and a sensitive galvanometer for measuring the small electromotive force set up in the thermocouple. Generally the apparatus also has a device for recording the changes.

The apparatus used in the detection of constitutional changes in metals has been discussed so frequently that further description is unnecessary here.[17]

The temperatures at which internal changes occur in a solid metal or alloy when heated or cooled are known as transformation points, or, as they are more commonly called, critical points. In iron and steel these critical points are given characteristic letters which identify them and fix their position. We will have occasion to look further at these characteristic points when we take up the iron-carbon diagram.

It often happens that the temperature at which a change takes place in the solid state in cooling does not coincide with the temperature for the same change when the material is heated. This variation which may amount to between 20 to 70 degrees Fahr. (10 to 40 degrees Cent.) is due primarily to lag or hysteresis. In other words the metal resists change when the change is due. When the change does occur it takes place with added violence. Probably the gap between the critical point on heating and the same point on cooling could be reduced to zero were the material heated and cooled infinitely slowly.

Practically, however, the gap between the two points always

[17] For a description of critical point apparatus the reader is referred to Rosenhain, "Introduction to Physical Metallurgy," Van Nostrand, 1917, p. 79 et seq. Sauveur, "Metallography and Heat Treatment of Iron and Steel," 1912, pp. 36 to 39; Williams, "Principles of Metallography," McGraw-Hill, 1920, p. 18 et seq. Hoyt, "Principles of Metallography," McGraw-Hill, 1920, p. 152 et seq. Heyn and Grossmann, "Physical Metallography," Wiley, 1925, p. 110 et seq.

exists and is taken into account in the heat treatment of metals, especially in steel. When all of the critical points are considered together we speak of the critical range, or, more properly, the transformation range.

In the construction of equilibrium diagrams the cooling curve is generally used, chiefly for the reason that it is easier to make an accurate cooling than heating curve. It is difficult to regulate the heating so that it will be absolutely uniform and result in a smooth curve. Irregularities due to uneven heating might thus be mistaken for heat effects.

Cooling Curves

There are four common methods of plotting the changes that occur when some metals and alloys are cooled from a high temperature. These are shown in Fig. 7.[18] The first and simplest is the time-temperature curve in which the time (t) is plotted as abscissæ and temperature (Θ) (in the diagram degrees centigrade) as the ordinates.

If a substance (neutral body) which does not undergo constitutional changes[19] is cooled from 1560 degrees Fahr. (850 degrees Cent.) and temperature readings taken at definite intervals, say 15 seconds apart; when the time (t) is plotted against (Θ) a curve similar to B will result. Now if the substance undergoes an internal change at 1525 degrees Fahr. (830 degrees Cent.) the heat evolved will be sufficient to hold the temperature almost constant for an appreciable time. When these results are plotted there will be a decided jog or flattening out of the curve as shown in A of Fig. 7.

After the transformation is complete cooling will again be proportional to the time until another change takes place, in this case, at about 1510 degrees Fahr. (820 degrees Cent.), where another jog will occur.

The time-temperature curve is satisfactory for the determination of freezing points, but in the case of small internal changes taking place in the solid metal it is not sufficiently sensitive. The data, however, procured by this method can be so interpreted that the change will be greatly exaggerated and thus made clearly evident.

[18] From Desch, "Metallography," Longmans, Green & Co., 1913, p. 127.

[19] For example; platinum, porcelain, steel containing 25 per cent nickel, etc.

The inverse rate curve (C, Fig. 7) is of this type. In this curve the ordinates are the temperatures (Θ). Instead of using the time as abscissæ, the length of time ($\triangle$t) required by the specimen to cool through a successive and constant interval of

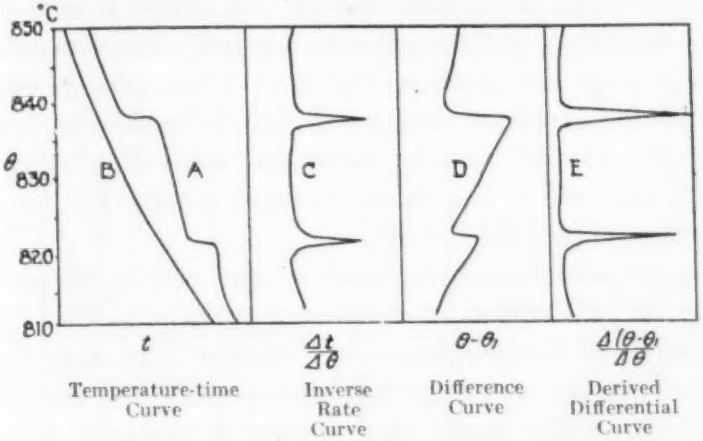

Fig. 7—Different Types of Cooling Curves (Desch).

temperature ($\triangle\Theta$) is used. In this type of curve, as long as the change in temperature is constant there is no abrupt change in the value $\frac{\triangle t}{\triangle \Theta}$ When the metal or alloy is undergoing an internal change heat is evolved and the temperature drop becomes very small; in fact it practically ceases. If the temperature Θ is plotted as ordinates and the relation $\frac{\triangle t}{\triangle \Theta}$ is plotted as abscissæ an evolution of heat is shown by a sharp break in the curve and a relatively long horizontal jog (C, Fig. 7).

The inverse rate curve is used extensively, especially where facilities are available for determining the time interval accurately.[20]

The time-temperature and inverse rate curves have the disadvantage that slight irregularities in the cooling of the furnace and other internal influences may introduce an error into the results. To overcome this the differential and derived-differential method of plotting may be used. These two methods produce a curve that is independent of any variations in the rate of heating or cooling.

In the differential method a substance known as a neutral body, which is free from constitutional changes, is placed in the

[20] A stop watch or chronograph is usually used.

furnace in close contact with the specimen. Two thermocouples are used. The specimen and neutral body are heated and cooled under identical conditions. As long as no internal changes take place in the specimen the millivoltage developed by both couples will be equal and the galvanometer will show a zero deflection. When the specimen undergoes even a slight constitutional change a relatively large deflection will occur. The curve is obtained by plotting the temperature Θ against the difference in temperature (millivoltage) $\Theta - \Theta'$ and is shown in curve D of Fig. 7. The differential method is the most common means for the accurate determination of critical points.

The derived-differential curve is the result of plotting the change in the difference in temperature $\triangle(\Theta - \Theta')$ for a constant interval of temperature $\triangle\Theta$ against the temperature Θ. This method serves to exaggerate the intensity of the heat effect at the transformation point and results in a curve similar to E of Fig. 7.

Apparatus for the accurate determination of the critical points in steel and other metals is now an essential part of practically every large scientific and steel works laboratory. In addition to the importance of an accurate determination of the position of these points in a scientific study of a series of alloys, each new alloy steel before it can be placed on the market must be investigated in order that the proper annealing and hardening temperatures will be known.

The Equilibrium Diagram

The equilibrium diagram is a graphic representation of the constitutional changes taking place in a series of metallic alloys. The complete diagram includes a thermal study of a large number of alloys. This thermal study consists of taking cooling curves from a point considerably above the beginning of solidification down to atmospheric temperature.

Fig. 8 shows one of the simplest forms of equilibrium diagram, the solidification of the alloys of cadmium and bismuth.[21] The first step in the study of these alloys is the determination of the freezing point of pure cadmium (K, Fig. 8). The curve of the pure metal shows uniform cooling until 609.6 degrees Fahr. (320 degrees Cent.) is reached. Here an evolution of heat occurs

[21]Diagram taken from Landolt, Börnstein, Roth, "Physikalisch-Chemische Tabellen," p. 658.

at which point the temperature is arrested and remains constant until all of the metal is solid. Cooling is then uniform until atmospheric temperature is reached.

Now consider an alloy containing 25 per cent bismuth and 75 per cent cadmium (M, Fig. 8). At 572 degrees Fahr. (300 degrees Cent.) the alloy is wholly molten. Cooling takes place at a uniform rate until a temperature of 500 degrees Fahr. (260 degrees Cent.) M^2, is reached, at which point an arrest occurs with evolution of heat. After this arrest the cooling takes place

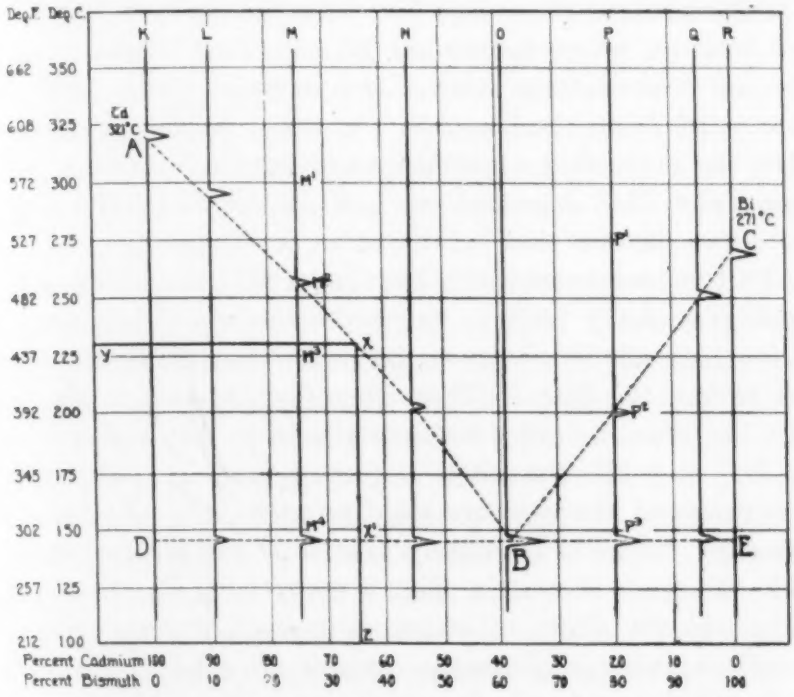

Fig. 8—Constitutional Diagram of the Cadmium-Bismuth Alloys.

uniformly until the temperature reaches 295 degrees Fahr. (146 degrees Cent.) M^4, where another evolution of heat occurs. After this, cooling will be uniform down to atmospheric temperature.

The alloy containing 61 per cent bismuth (O, Fig. 8) cools uniformly until 295 degrees Fahr. (146 degrees Cent.) is reached; this alloy shows only one arrest.

In the diagram shown in Fig. 8 the arrest points for only five alloys are plotted. Actually in constructing an equilibrium diagram a large number of alloys with varying percentages of the two metals are studied.

In constructing the diagram the line ABC is drawn through

the upper points and the line DBE is drawn through the lower points.

After the diagram is constructed the arrest points for a cadmium-bismuth alloy of any composition may be determined. For example, we have an alloy containing 35 per cent bismuth, an ordinate run up from 35 per cent (Z) will intersect the line AB at X. The horizontal line XY from this point will tell us that the first evolution of heat for the alloy containing 35 per cent bismuth occurs at 446 degrees Fahr. (230 degrees Cent.).

If the process of cooling has been closely watched it will be found that an alloy containing 35 per cent bismuth is wholly molten at temperatures above 446 degrees Fahr. (230 degrees Cent.), point X on the line AB. At point X the jog in the curve denotes the beginning of solidification which continues until 295 degrees Fahr. (146 degrees Cent.), X^1, is reached. Here the break in the curve denotes that solidification is complete.

At all temperatures above the line ABC the alloys of cadmium-bismuth are wholly molten, hence the line ABC is known as the *Liquidus*. At all temperatures below the line DBE the alloys are wholly solid; the line DBE is, therefore, known as the *Solidus*.

In the area included between the lines AB and BD and between BC and BE the alloy is partly molten and partly solid. In this range of temperature solid particles are suspended in the molten alloy. This is familiarly known as the mushy stage.

The diagram shown in Fig. 8 represents simple equilibrium conditions in the alloys of cadmium-bismuth during solidification. When alloys undergo allotropic changes after solidification is complete, as is the case in the iron-carbon or steel series, the diagram becomes much more complicated. However, the construction of the diagram involves the same procedure; namely, the accurate determination of the various points where heat is evolved in cooling from the molten to atmospheric temperature, followed by the plotting of these temperatures for a large number of alloys of the same series.

Interpretation of the Equilibrium Diagram

Referring to Fig. 8 we note that pure cadmium (K, Fig. 8) freezes (solidifies) at 609.6 degrees Fahr. (320 degrees Cent.). There is only one arrest on the cooling curve of the metal. As we have noted this is a characteristic of pure metals; that they

have a definite and constant freezing (melting) point. In the case of the alloy containing 25 per cent bismuth (M, Fig. 8) two arrests occur on the cooling curve; the first at 500 degrees Fahr. (260 degrees Cent.) and the second at 295 degrees Fahr. (146 degrees Cent.). Referring again to Fig. 8, we see that the alloy containing 45 per cent bismuth (N) has an arrest at 392 degrees Fahr. (200 degrees Cent.) and one at 295 degrees Fahr. (146 degrees Cent.), and the alloy containing 61 per cent bismuth (O) has only one arrest on this curve, at 295 degrees Fahr. (146 degrees Cent.).

Fig. 8 shows us that with the exception of the one containing 61 per cent bismuth, none of the alloys of cadmium-bismuth have a constant and definite freezing point, like a pure metal. On the contrary they have a solidification range with a definite temperature where solidification begins and a definite temperature when it is complete.

We shall see later that alloys similar to the 61 per cent bismuth-cadmium alloy do have a definite freezing point. We shall also see that in many alloy series there is usually one alloy which is of definite composition and which has a definite freezing point, generally lower than the rest of the series.[22]

This alloy of definite composition which freezes at a definite temperature is known as the *eutectic* alloy. An understanding of the characteristics of eutectic alloys is of vital importance to a clear comprehension of the constitutional changes in iron and steel. The eutectic or alloy of lowest melting point enters importantly into the constitution of cast iron. In steel there are changes taking place in the solid state which give rise to a constituent that closely resembles a eutectic alloy in its properties and appearance.

Referring again to Fig. 8 we see that with increasing percentages of bismuth (from 0 to 61 per cent) the temperature of the first (upper) arrest in the cooling curve gradually becomes lower, along the line AB. It is also evident that the temperature of the second arrest at 295 degrees Fahr. (146 degrees Cent.) remains constant for all of the alloys of these two metals.

Having obtained this experimentally from the cooling curve determinations let us attempt to interpret the data. The first

[22]This is not true in the case of solid solutions, i. e., a solution of one metal in another in the solid state. This will be discussed in the next installment.

conclusion forced upon us is that introducing molten bismuth to molten cadmium lowers the temperature at which solidification begins. Another factor enters with the introduction of the second metal, the appearance of the second arrest at the temperature 295 degrees Fahr. (146 degrees Cent.). As the percentage of bismuth increases (from 0 to 61 per cent) the upper arrest is lowered progressively until it coincides with the lower arrest at the point B. With further increasing percentages of bismuth, the upper point again appears and rises progressively along the line BC until when the bismuth content reaches 100 per cent the upper arrest is at 500 degrees Fahr. (271 degrees Cent.) and the lower arrest has disappeared.

If the cooling curves obtained on a series of these alloys are studied it will be noted that the intensity of the upper arrest which is great for pure cadmium decreases until it can scarcely be detected when the bismuth is close to 61 per cent. Likewise the intensity of the lower point increases from zero for pure cadmium to the maximum at 61 per cent bismuth. The same holds true for alloys containing more than 61 per cent bismuth, the intensity of the upper point increases along the line BC as the bismuth percentage increases while the intensity of the lower point decreases along the line BE as the percentage of cadmium decreases.[23]

In our study of the theory underlying the constitution of iron and steel we will encounter similar arrest points denoting the beginning and end of solidification. In addition we will encounter arrest points which become evident when cooling curves are taken of iron and steel after it has completely solidified. We will find that these latter arrest points change in position with a change in carbon percentage.

If we study a simple alloy such as cadmium-bismuth in detail; when we come to a discussion of the iron-carbon diagram the theoretical considerations underlying the constitutional changes in iron and steel when it is heated and cooled will become clear and relatively simple.

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[23]We will see later that this increase in the intensity of the constitutional change is especially evident in the changes taking place in solid alloys.

EDITOR'S NOTE: The next installment will discuss different classes of metallic alloys, including the constitution of eutectics and solid solutions.

MULTIPLE BREAKING

By W. R. Ward

Abstract

This paper describes a rapid and efficient method of breaking bars into multiple lengths for use in forging.

A method is described whereby the bars are notched alternately 180 degrees apart into required billet lengths, by means of an oxyacetylene torch. By applying simultaneously a number of alternate plus and minus bending moments the pressure of the fulcrum causes the bar to break as shown in Fig. 1.

The method of notching on one side of the bar only is likewise described and the impracticability of this method is shown.

IN breaking steel by notching with a torch and applying a force to break the piece, it is essential to apply the force so that it will cause a bending moment that tends to open up the notch or "nick." As the steel is weakened at this point by the reduced section and also made somewhat brittle by the steel being hardened from the chilling effect of the cold metal around the highly heated nick, it breaks easily when strains are concentrated at this point. Steel low in carbon, however, does not break as well as higher carbon steels.

During the world war the method was used extensively in making billets for various shell and other forgings: presses were used instead of rams or hammers to exert the necessary force to break the steel that had been nicked with a torch. Various types of saws and cut-off lathes were used to form the billets previous to using a press to break them.

In conjunction with some of the war work, the writer was asked to arrange a press to break billets, one at a time, for each stroke of the press as has been done in other shops. However, as the number of presses available for this purpose was limited, and as the maximum production had to be obtained from each press, the writer conceived the idea of breaking several billets per stroke of the press. To do this, however, required a different

The author, W. R. Ward, Lyells, Virginia, is a member of the Lehigh Valley Chapter of the Society.

method of nicking the bars than had hitherto been used; they were alternately nicked first on one side and then the other throughout the length of the billet: that is the double length on one side. (Fig. 1). This shows how the forces are applied. The bending moments are plus and minus alternately. It will also be seen from

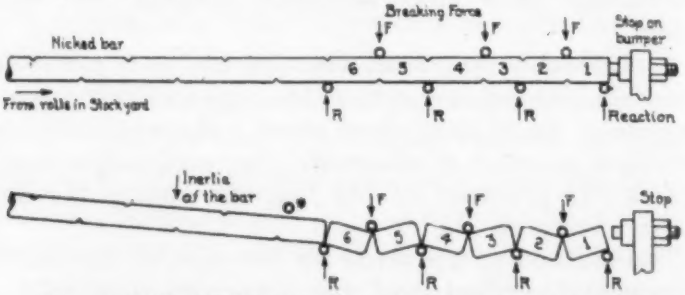


Fig. 1—Nicked Billets or Bars Showing Position of Fulcrums Before and After Breaking.

Fig. 2, that a bar nicked on one side would be difficult to break in a number of pieces at one time, as the bending moments are all in one direction. If forces could be applied evenly and all the billets broken at once on the one side nicked bar, it would tend to curl down as shown. An impractical operation.

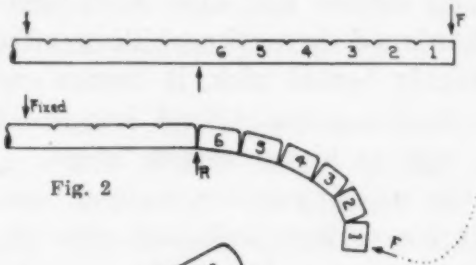


Fig. 2

Fig. 3

Fig. 2—Showing a Bar with the Nicks on One Side and the Manner in which it Breaks. Fig. 3—Shows a Marking Gage Used in Marking Position of the Notches.

A gage as shown in Fig. 3 was made to lay off the bars that were spread out in one layer on I-beam racks (Fig. 4) using soap-stone to mark the bars. The bars were first nicked on one side double billet length and then rolled 180-degrees and nicked on the other side, with a torch.

A 500-ton press was first fitted with two large I-beams of sufficient length to hold several breaking fulcrums. The beams were found to be too flexible in the webs to stand the strains,

however they broke enough billets to demonstrate that the idea was sound. A 200-ton press was then fitted with a multiple breaking device designed by the writer that was made of cast steel and strongly braced. (Fig. 5). This worked satisfactorily and produced 149 billets 3⅜ inches in diameter by 4½ inches long in 20 minutes when first tried; later the press was run by three men (not counting the crane man who took the billets away) who started at 7:30 a. m. and stopped at 2:55 p. m. (25 minutes out for lunch), they broke 3400 billets, the same size as above. The pho-

Fig. 4—Showing the Method of Nicking Bars with the Acetylene Torch.

tograph shows five of them broken at one stroke. Sometimes six billets would be broken when the bars were long (the inertia of the bar breaking the sixth), then again less than five would be broken, but five was the average number.

It would be possible to break more than one bar at a time by having wider breaking fulcrums.

The press shown handled the work that was done on many cut-off saws.

Forged fulcrums would wear out under such severe service, and it was a question of renewing them after short use. The problem of producing the simplest and cheapest fulcrum was paramount. Finally a short length of medium high carbon round hot-rolled steel cut to length with a torch was adopted. They were reversible or in fact had a number of positions they could

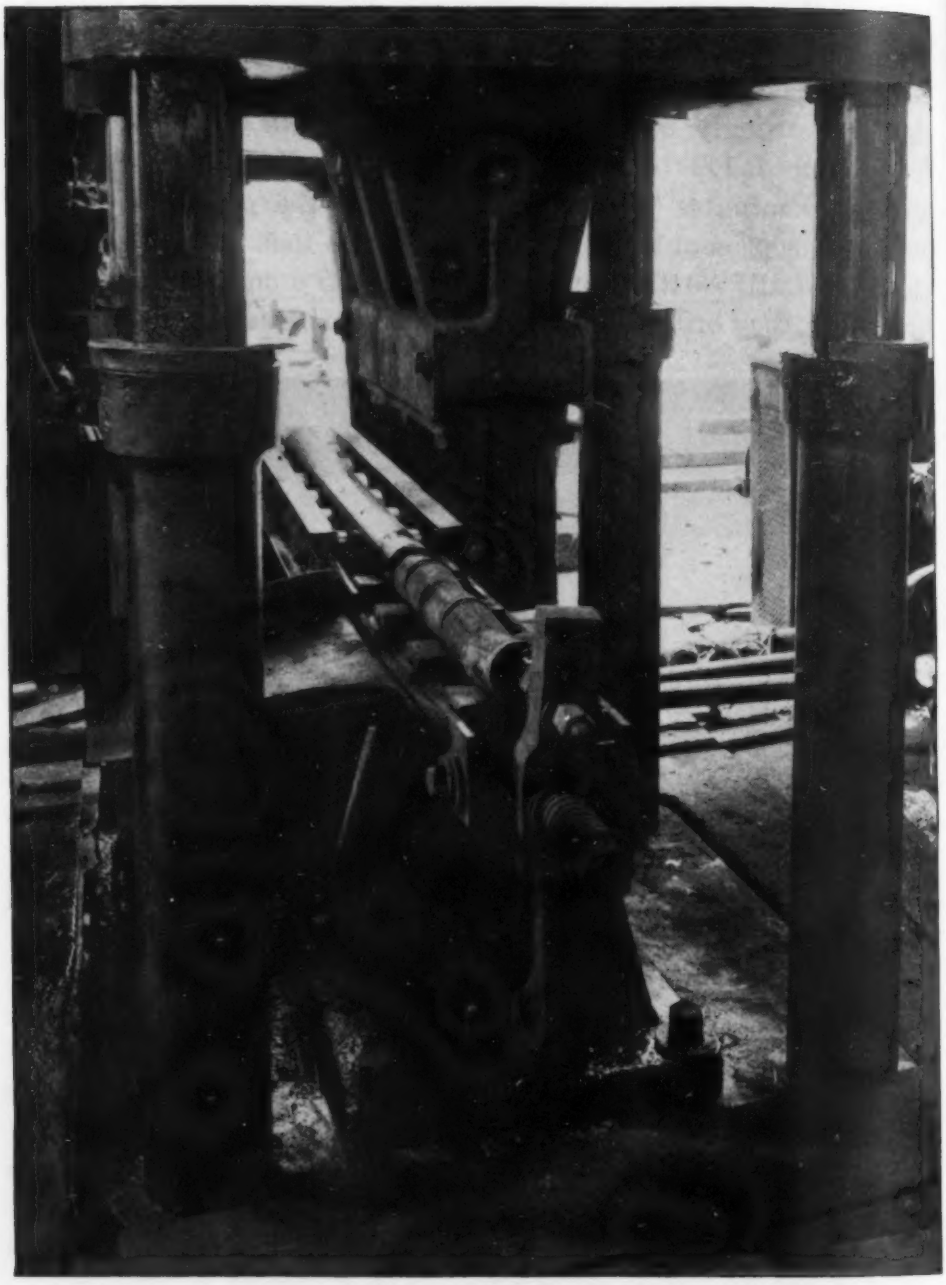


Fig. 5—Showing a 200-ton Hydraulic Press Equipped for Breaking Bars Into Billet Lengths.

turn to the bars as indications of wear appeared. The sketch of Fig. 6 shows how the fulcrums were supported.

It will be noted from Fig. 5, that a bumper is located at the end of the breaking device. This was found necessary as the nicked bars were fed by electrically operated rolls placed outside

of the building in the stock yard, to the breaking press, and the motion of the bars had to be arrested suddenly when the adjust-

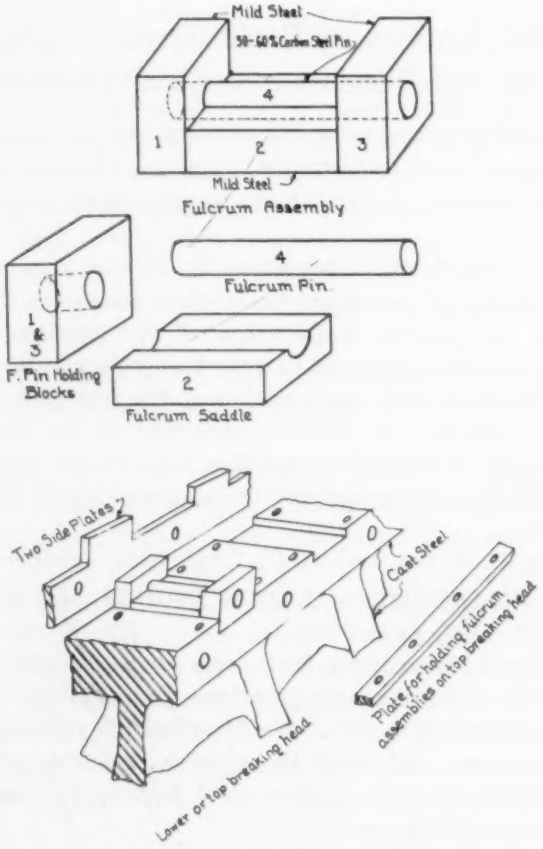


Fig. 6—Sketches Showing the Details of the Design of the Breaking Press Head.

ing stop screw on the bumper came in contact with them. This screw gaged the nicks so they were opposite the fulcrums. The operators had to see that the bars were placed with nicks in the proper position.

Abstracts of Technical Articles

Brief Reviews of Publications of Interest to Metallurgists and Steel Treaters

ON THE HEAT TREATING AT THE DRAWING OF THE STEEL WIRE. By S. Steinberg, *Journal* of the Russian Metallurgical Society, 1925, I, 131.

Considering the mechanical characteristics of the steel wire, obtained by tempering and subsequent drawing, the author concludes that the metal of the wire shows a remarkable combination of the qualities of strength and tenacity; a similar combination can be stated only in special steels after their heat treating. The mentioned amelioration of the qualities of carbon steel is explained by two reasons—the sorbitic structure of the metal and its cold working. The author forwards the question were it not possible to use more widely the advantages of the mentioned treating and apply it to larger articles —rails, beams, etc.

Abstracted by M. Oknoff and G. N. Koulginski, Leningrad, Russia.

CHEMICAL ANALYSES OF LARGE INGOTS. By J. H. Hruska. In *Iron Age,* April 15, 1926, page 1049.

The author is of the opinion that facts as to inclosures and gases are more vital than strict chemical specifications and that melting practice and analysis are both important factors. He describes the methods of manufacture by the various processes, and is of the opinion that the greatest number of large ingots, particularly plain carbon steel ingots, are made mostly from basic, or sometimes, acid, heats.

ON THE RELATION BETWEEN THE LATTICE-CONSTANT AND THE DENSITY OF IRON-NICKEL ALLOYS. By Atomi Osawa. Preprint, British Iron and Steel Institute, May, 1926.

By experiments, the author determined for a whole series of iron-nickel alloys the lattice constants and the densities for both annealed samples and those dipped in liquid air. The writer by means of photomicrographs, tables and diagrams adds to the value of his paper and a short summary sets forth his findings.

THE ESTIMATION OF PHOSPHORUS IN STEELS CONTAINING TUNGSTEN. By T. E. Rooney and L. M. Clark. *British Iron and Steel Institute,* May 1926.

The authors carried out a large number of experiments on steels of various tungsten content and give their findings in this paper. They incorporate a description of their experimental work and by various tables show the results obtained with a series of samples prepared in various ways.

THE INFLUENCE OF SEGREGATION ON THE CORROSION OF BOILER TUBES AND SUPERHEATERS. By G. R. Woodvine and A. L. Roberts. *British Iron and Steel Institute,* May 1926.

The authors are of the opinion that rapid steaming, due to the use of high-pressure water-tube boilers and superheaters, which unquestionably has led to a saving in fuel, has also brought forth questions relating to boiler-making materials. They set forth that not much improvement has been made in the mild steel vessels in which the treated feed-water is converted into steam although valuable observations have been made in the treatment of feed-water. They further set forth that it is their belief that the steel tubes should be made from ingot material of the non-piping variety, as results observed by using tubes made of sound steel have been most satisfactory.

The authors have established by observation and tests described in this paper, their belief in the connection between segregated ingot material and defects in such tubes. It may be, they say, that other tube failures, such as domestic hot-water systems and upwards, are to some extent attributable to these causes.

TESTINGS OF MATERIAL IN THE PRACTICE. By Dr.-Ing. M. Moser, Essen. *Zeitschrift des Vereines Deutscher Ingenieure,* March 13, 1926, pages 341 to 347.

The author shows the development in the testing of materials, the change from direct to indirect methods, which still have many imperfections.

Further he shows the influence of testing-conditions and irregularities of materials.

Then the author gives some details such as: the determination of the yield-point, the interpretation of the appearance of fracture of tensile—and bending tests, the testing of hardness by the Brinell-method, the dynamical hardness-testers and by the scleroscope, the bending test, the shock test, the endurance-shock test, etc.; further the selection of testpieces of a turbo-dynamo magnet wheel, the examination of a railway-coupling, etc.

Finally the author discusses the need for absolute reliability of testing methods. *Abstracted by Dr. Hans Pollack.*

SMELTING SECONDARY ALUMINUM AND ALUMINUM ALLOYS. By Dr. R. J. Anderson. In *Metal Industry,* May 1926, page 187.

This is Part IV of a series of articles on the reclamation of all forms of scrap and used aluminum and aluminum alloys and deals primarily with the fluxes for use in secondary smelting.

The author sets forth that fluxes are used to separate the dross from the metal and to prevent oxidation. He claims that the different fluxes vary depending upon their chemical composition and he divides fluxes for secondary aluminum smelting into four main classes. These he discusses at some length. He also sets forth the requirements of a smelting flux.

He gives a list and detailed description of fluxes used in aluminum and alloy smelting practice such as: aluminum chloride, binary salt mixtures, borax, calcium chloride, complex salt mixtures, cryolite, fluorspar, sodium chloride, welding fluxes, and others.

THE COLD DRAWING OF BAR STEEL. By F. W. Krebs. In *Mechanical Engineering,* May 1926, page 448.

In this article, the author sets forth the object of cold finishing and enumerates the four purposes for which cold drawing or cold rolling may be

employed. The pickling process is described, the cleansing of the bars and the pointing. He then describes the dies used and also the drawing of the steel, after which he takes up the manner of finishing the bars. The factors affecting the machining qualities are considered and the writer is of the opinion that grain structure and hardness largely determine the machinability of steel. The physical properties of several steels and comparative results between various methods of processing are tabulated.

STANDARD PRACTICE FOR MOLDING. By Paul R. Ramp. In *Iron Age*, March 18, 1926, page 753.

When the best process of molding certain pieces or kinds of work has been decided upon, the adoption of a standard practice is of economic importance, is the opinion of the author of this article.

Definite instructions to the foreman and to the workmen produce better results and less waste. In shops producing large castings of different designs the establishment of molding practices is of necessity slow. But the unconscious application of the standard practice methods to other kinds of work will be valuable.

The writer continues with a detailed description of the standard practice for molding the stern tube in a ship. Twelve illustrations are given in this connection.

The adoption of the described standard practice has resulted in a casting of higher grade, the cost of molding is reduced 50 per cent and apprentices can do the work required to be done.

ON GHOST LINES AND THE BANDED STRUCTURE OF ROLLED AND FORCED MILD STEELS. By J. H. Whiteley. *British Iron and Steel Institute*, May 1926.

The authors are of the opinion that only when variations in the percentage of phosphorus between two adjacent areas in iron exceed 0.07 per cent do they cause the removal of carbon, between Ar3 and Ar1, from the richer area.

They also give evidence that in certain cases carbon may move from one region to another of higher phosphorus concentration. They also give consideration to the theory that ghost lines are due to crystallization of ferrite on non-metallic inclusions and point out that they cannot adequately be accounted for in that way.

THE EFFECT OF MASS IN THE HEAT TREATMENT OF NICKEL STEEL. By Walter Rosenhain, R. G. Batson and N. P. Tucker. *British Iron and Steel Institute*, May 1926.

The authors set forth that it is a well-known fact that in the heat treatment of steel the size of the piece has an effect on the properties obtained. Mass has an influence on heat treatment as evidenced by the failure in service of several pinions of relatively large size from the double reduction gears of marine steam-turbines. They set forth the results of their investigation to ascertain the cause. Both the mechanical tests and the microscopical examinations are given in some detail and the effect of heat treatment on the structure and properties of the steel is set forth.

MARKED PROGRESS IN MACHINE FORGING. By C. D. Harmon. In *Forging-Stamping-Heat Treating*, April 1926, page 122.

The author makes a comparison of machine forgings and drop forging and is of the opinion that the quality of machine forgings has been improved more than is generally appreciated. He sets forth in his paper some of the features which have advanced the art of machine forging to its present position. He further states that whereas the forging machine was in the past a crude tool, it has been developed and perfected until it now produces machine forgings of such quality that it has become a close rival of the drop hammer which has been for many years an efficient tool.

The writer explains the competition of the forging machine and the drop hammer and the operations of each, and is of the opinion that when work can be made by either method, the forging machine cost is lower and states his reasons for believing this to be the fact.

By means of photographs and diagrams he adds to the concreteness and interest of his subject.

ON THE MACROSCOPIC DETERMINATION OF CARBON CONTENT OF SAMPLES IN OPEN-HEARTH STEEL MANUFACTURING PROCESS. By Matsujiro Hamasumi. *Kinzoku no Kenku* 3, 1926, No. 1, pages 54-64.

It is an important part of the daily work of steel melters in the open-hearth steel making process to estimate the carbon content of the steel bath macroscopically by the fracture of quenched steel samples; the ability of this estimation is the result of many years' experience of steel melters.

In this article, the writer explains metallographically the characteristic nature of the fracture of quenched steel samples by the measurement of hardness and angles of bending by blow, combined with the change of microstructures in these samples. He attributes this difference of fractures for different carbon content to the inner structure caused by the relation of the lowering of the A_3 point due to carbon content to constant quenching temperature. The writer also explains how the presence of manganese influences the fracture of basic steel samples on the basis of the equilibrium diagram of the ternary system iron-carbon-manganese. *Abstracted by Dr. Kotaro Honda.*

MAKING HIGH GRADE STEEL—XI. By John A. Coyle. In *Iron Trade Review*, May 6, 1926, page 1177.

In this paper the author sets forth the wide range of analyses, the many degrees of hardness, and the rolling of unusual shapes required for tools. In a table he enumerates the special steels for tools by various processes and states that originally all of the special steels listed in the table were produced in the crucible and he is of the opinion that for steels which must maintain a good cutting edge and give the maximum of wear, crucible steel is preferred.

By means of illustrations, he shows some of the special shapes turned out by the mills such as dies, tools, drills, bars, axes, skates, etc., and points out that coordination of the user and maker will result in improved products at less cost.

THE THEORY OF CAST IRON. By Prof. Dr. H. Hanemann. V. D. I. Nachrichten, March 24, 1926.

The mechanical properties of cast iron are dependent on the nature of the metallic matrix and on that of the segregation of graphite. The matrix consists either of ferrite, of ferrite and pearlite or only of pearlite; the

graphite may be coarse-flaked, fine-flaked or eutectic. The tensile strength of cast iron with coarse-flaked graphite amounts to 14,000 to 20,000 pounds per square inch, with fine-flaked graphite 28,000 to 34,000 pounds per square inch and with eutectic graphite 43,000 to 50,000 pounds per square inch.

The crystallization of graphite takes place about 2000 degrees Fahr. during the solidification. The matrix, being formed at the same time with it, consists at first of a solid solution of iron and carbon. In the range from 1830 to 1300 degrees Fahr. the permanent state is formed by transformations in the solid state.

On further cooling the above-mentioned solid solution changes its percentage of carbon according to the diagram of the system iron-carbon and the percentage of carbon decreases. If the cooling would take place in a very slow manner the total amount of carbon would be separated as graphite. But usually the crystallization leaves the iron-carbon system in some point and follows the system iron—iron-carbide. Changing the cooling conditions it is possible to obtain either a ferritic-pearlite or only a pearlitic matrix. The amount of silicon is of great influence.

Melting cast iron, only a part of the graphite is dissolved; the rest is preserved as graphite in the fusion for a time. Consequently there is in the fusion a great number of graphite nuclei and when cooling begins, the graphite crystallizes on these nuclei. It results in a so-called pseudo-eutectic mixture, that is the flaked graphite.

But if the graphite nuclei are destroyed the crystallization takes place in such a manner, that at first mixed crystals crystallize out and referring to that the real eutectic mixture, the fine-grained, eutectic graphite.

This interpretation is confirmed by experiments. Heating the molten cast iron to 2730 degrees Fahr. destroys the graphite germs and mixed crystals and eutectic graphite crystallize out. The crystallization of the latter is independent of the rate of cooling.

The experiments were verified in the practice and it was confirmed that it is possible to manufacture castings of any size, applying this theory.

The moulding and the technological treatment offer no special difficulties.

Abstracted by Dr. Hans Pollack.

PLATING STEEL AUTO PARTS WITH CHROMIUM. In *Iron Trade Review*, May 6, 1926, page 1182.

Chromium plated steel is now supplanting nickel plated steel for various purposes in the industrial world. While the process of chromium plating has been known for some time, it was too expensive to be used extensively commercially. This drawback to its universal use, especially in the automotive industry, has now been eliminated. Chromium plated radiators are now being used by many automobile concerns. Lamp shells, bumpers, crank hole caps, motormeters and hub caps are now also being chromium plated.

This new process necessitates factory changes and is a purely plating method. There is no change in the original surface. Iron and steel are first given a coating of copper and nickel and then a deposit of chromium. The result is a lustrous silvery-white. This coating is more rust-resisting than heavy nickel.

In the chromium plating process there is a saving in labor cost which offsets the added cost of material, if any.

The solution used consists of chromic acid, chromium sulphate and boric acid.

The article sets forth the equipment necessary and also describes the method of cleaning the parts.

The supply of chromium is said to be practically inexhaustible, in fact, it is so great that it is used in road maintenance work.

Its extreme hardness enhances its wearing properties and experiments are constantly being made with a view to using it possibly for crankshafts, valves, tappets and other automobile parts.

OCCLUSION OF GASES BY METALS AND ALLOYS IN LIQUID AND SOLID PHASES. By Keizô Iwasé. In *Kinzoku no Kenku* No. 3, 1926, page 119.

The writer measured the quantity of different gases which can be absorbed or evolved by metals at high temperatures without causing any chemical reactions. The method of measurement was that of A. Sieverts improved by the present writer. As the quantity of gases absorbed, the saturation values of the quantity of these gases, which a metal can absorb, were taken. The results are principally given in curves. *Abstracted by Dr. Kotaro Honda.*

DETERMINATION OF THE DENSITY OF CEMENTITE. By Toyozô Ishigaki. *Kinzoku no Kenku*, No. 3, 1926, page 169.

The writer determined the density of cementite by measuring steels of different carbon contents, by applying suitable factors for other impurities than carbon and by extrapolating to the concentration of cementite. The result was 7.657 at 4 degrees Cent. *Abstracted by Dr. Kotaro Honda.*

INVESTIGATION OF THE SPINDLES FOR THE TEXTILE MACHINES. By G. L. Sakharoff. *Messenger* of the metal Industry (Moscow), 1925, I-II, 126-133.

After having investigated the present spindles chemically and metallographically the author comes to the conclusion that the most suitable for this purpose appears to be the basic Martin steel of the following composition:

	per cent
Carbon	1.0 –1.1
Silicon	0.20–0.25
Manganese	0.55–0.65
Phosphorus	less than 0.03
Sulphur.....	less than 0.03

Rolled and cut billets of the steel must be subjected to annealing before turning in order to get a more homogeneous structure. The hardening of the spindle after the mechanical working must be done so that its body had a troostite-sorbite structure, and the sharp end, an austenite-martensite one. For this purpose, the spindle, heated up to 800-850 degrees Cent., is hardened by dipping into oil and after this the sharp end is again heated up to 750 degrees Cent. and hardened in cold water. For removing the detrimental stresses

the sharp end must be subjected to tempering at 200 degrees Cent.

Abstracted by M. Oknoff and G. N. Koulginski, Leningrad, Russia.

PROPERTIES OF HIGH SILICON STEEL. *Stahl und Eisen*, April 15, 1926, p. 493-503.

It is known, that an addition of about 1 per cent silicon to mild carbon steel increases the tensile strength and the yield point, without impairing the elongation.

Recently the Berliner A.-G. für Eisengiesserei und Maschinenfabrikation, früher J. C. Freund & Co., Berlin, proposed the use of high silicon steel as structural steel, and through this the question of the utilization of the above mentioned influence of silicon was put anew.

By numerous tensile tests on rolled bars of low carbon content and about 1 per cent silicon melted in the Bosshardt furnace, the influence of such a silicon content, already determined by previous examinations, was confirmed. The tensile strength and the yield point increased without noticeable decreasing of elongation and contraction.

The tensile properties, peculiarly the yield point, depends on the profile dimensions respectively on the rolling conditions.

Heats of the same chemical composition, manufactured in the ordinary open-hearth furnace, in the Thomas converter or in the electric melting furnace, proved it possible to produce in these furnaces a steel of the same tensile properties as in the Bosshardt furnace, although the high qualities of this new silicon steel were attributed by the Freund A.-G. to the melting in the Bosshardt furnace. *Abstracted by Dr. Hans Pollack, Germany.*

METHOD OF SHRINKAGELESS CASTING OF CRUCIBLE STEEL AND OF MARTIN STEEL. By N. Grammatchikoff. *Messenger* of the metal Industry (Moscow), 1925, III-IV, 14-21.

A method of making steel castings is described which is giving good results at one of the Russian works. The method consists at first in using a mold widening to the top with a spherical bottom and a pyramidal lined cover above. The solidification of the metal in such a mold takes place by layers from below, while the upper part of the liquid metal feeds continually the funnels being formed during the process of solidification. The last funnel above is filled with liquid metal from the cover. The mold widening at the top contributes also to the easy evolution of gases from the liquid steel, and there are therefore produced ingots without blisters. Special attention is devoted to the reduction of the metal. It is being done in a bath with usual reducers (FeMn, FeSi, etc.). When in ladle it is achieved with a small quantity of charcoal, thrown into the ladle before the admission of metal, and finally when in a mold the reduction is attained by aluminum, thrown in in small pieces, forming 0.004-0.009 per cent of the weight of the steel. In the above described method of casting the percentage of waste for the different kinds of steel does not exceed 10-12 per cent.

Abstracted by M. Oknoff and G. N. Koulginski, Leningrad, Russia.

REFRACTORY MATERIALS IN THE IRON AND METAL INDUSTRY. By Dr.-Ing. E. H. Schulz, Dortmund. Zeitschrift des Vereines Deutscher Ingenieure, March 20, 1926, p. 408-410.

In this paper the author shows a way for investigation, testing and standardizing of refractory materials, on which great demands of physical and chemical nature are made by the consumers.

He describes briefly the principal sorts of refractory materials and their properties, important for the behavior in working, as chemical composition, refractoriness, changes of volume and of shape, resistance to chemical corroding agents as slags, flue dust, ashes and porosity.

Of the testing methods, being at our disposal, the determination of the chemical composition and of the specific gravity are comparatively simple. The determination of the melting point by Seger cones is replaced by the softening test under pressure at high temperature. Another testing method is the examination of shrinking and expansion. The examination of the behavior against sudden change of temperature and that of resistance against corroding agents encounter great difficulties.

Finally the significant investigations of E. Steinhoff are mentioned, in which he attempted to examine the structure of refractory materials by etching of slides. *Abstracted by Dr. Hans Pollack, Germany.*

WEAR TESTING OF MATERIALS FOR RAILWAYS AND INDUSTRIAL WORKS. By Oberbaurat Ing. Spindel, Innsbruck. Zeitschrift des Vereines Deutscher Ingenieure, March 20, 1926, p. 415-421.

The wear testing machine of the Maschinenfabrik Augsburg-Nürnberg, designed by the author, is fully described. It serves for testing the wear-resistance of all sorts of structural materials, of tools and tool steels, lubricants, etc. The examination of ordinary rail steel is carried out with a grinding velocity of 30-35 meters per minute (1200 to 1400 inches) and a load of 5 kilograms (11 pounds), wear resistant steels are tested with a velocity of 20-25 meters per minute (790 to 985 inches) and a load of 10 kilograms (22 pounds). The testing duration amounts 3-15 minutes.

The gliding friction is examined by pressing the test piece against a rotating disk of 1 millimeter in thickness, that grinds out a segment-formed cut from the test piece; by this the disk is also worn. The ratio of the volume, removed by grinding, to the grinding time or the grinding way is the measure for the wear of the material; the ratio of the volume, worn out of the disk, to the grinding way is the measure for the reaction of the test piece on the grinding disk.

The inverse values of these ratios are numbers of quality for the wear-resistance and the saving-resistance, from both of these numbers is concluded the real wear-quality of the material.

The results, obtained by examination of open-hearth and Thomas steel rails with tensile strengths from 85,000 to 128,000 pounds per square inch, show that the wear-resistance is independent of the tensile strength.

The question of wear-quality may only be solved with the aid of the metal-microscope and the wear testing-machine.

Abstracted by Dr. Hans Pollack, Germany.

AN INVESTIGATION ON BLACK HEART MALLEABLE CAST

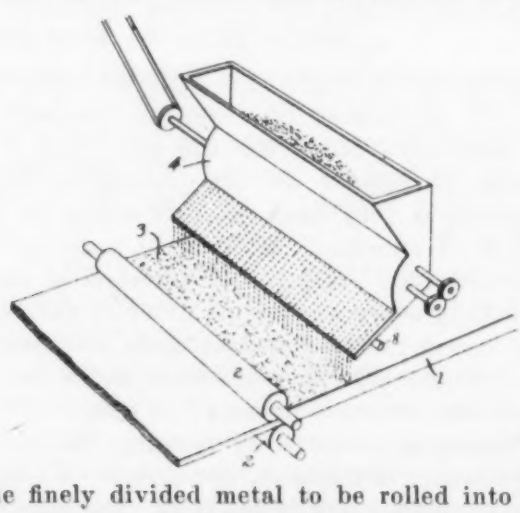
(Continued on Page 138)

Reviews of Recent Patents

By
NELSON LITTELL, Patent Attorney
475 Fifth Ave., New York City
Member of A. S. S. T.

1,580,647, Process for Treating Ferrous Metal to Prevent Corrosion, George D. Breck, of Cleveland.

This patent describes an apparatus and process for treating ferrous metal products to render the same resistant to corrosion. The process comprises rolling a finely divided corrosion resisting metal capable of forming an alloy with the iron or steel into the hot surface of the iron or steel during the rolling operation. The apparatus comprises a hopper 4



for containing the finely divided metal to be rolled into the surface, such as chromium, or a mixture of chromium and iron, which hopper is provided with suitable means for uniformly feeding the powdered mixture onto the surface of the plate 1 which is being rolled. The powdered material 3 is distributed over the surface of the plate just prior to the passage of the plate through the rolls 2 and on the passage of the heated plates through these rolls, the surface coating of chromium or the like is rolled into the surface of the sheet, the heat of the sheet and the pressure of the rolls combining to form an alloy coating on the surface of the sheet. If desired, the powdered coating material may be heated as it passes over the apron of the hopper 4 by means of a gas burner or the like 8.

1,580,662, Alloy, Pierre Girin, of Paris, France, Assignor to Societe Anonyme De Commentry Fourchambault and Decazeville, of Paris, Seine, France.

This patent describes an alloy for use particularly at extremely low temperatures, such as are encountered in the liquefication of permanent

gases, and which will be free from the brittleness of ordinary iron and steel at these temperatures and will have a high elastic limit and tensile strength. The alloy is particularly adapted for use in the manufacture of apparatus for liquefying air, nitrogen, or the like, at temperatures of minus 180 degrees Cent. or below. The alloy may contain nickel 40 to 70 per cent, manganese 0.8 to 5 per cent and the remainder iron. The addition of small percentages of vanadium and titanium, that is 0.1 to 0.5 per cent, has a favorable action on the quality of the alloy. A preferred composition of the alloy is nickel 55 per cent, manganese 2 per cent and iron 43 per cent.

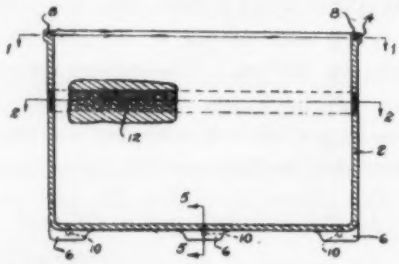
1,581,269, Process of Treating Metals to Inhibit Excessive Grain Growth, George L. Kelley, of Philadelphia, Pennsylvania, Assignor to Edward G. Budd Manufacturing Company, of Philadelphia, Pennsylvania, a corporation of Pennsylvania.

This patent describes the method of prohibiting excessive grain growth, especially during the annealing of sheets or the like which have been subjected to a cold working operation, such as drawing. The inventor has noticed that carbon steel, when subjected to cold working and then to an annealing temperature, develops abnormally large crystals which reduce the strength of the material. He has found a method of overcoming this tendency which comprises subjecting the articles, which have been put in a condition of strain through cold working to a preliminary heating, which is below that at which excessive grain growth develops, for a sufficient length of time to relieve the strain of cold working, after which the article may be heated to the annealing temperature without the development of the large grain growth. As one example of the process, he heats a low carbon steel of 0.05 to 0.15 per cent carbon, which has been subjected to a cold working operation, to a temperature of approximately 1100 degrees Fahr. and maintains the article at this temperature for a period of about two hours to relieve the working strain. After this period, the temperature is increased to 1200 to 1400 degrees Fahr. for the purpose of annealing the article, and no excessive crystal or grain growth will be developed.

1,581,443, Heat-Resisting Article, John Clifton, of New York, N. Y., Assignor to Driver-Harris Company, of Harrison, New Jersey, a corporation of New Jersey.

This patent describes the production of heat-resisting boxes, such as annealing boxes, retorts and other castings which are to undergo repeated heating or cooling operations. The article comprises a main body portion, such as an annealing box 2, provided with feet or supports 6, upon which the box rests and having imbedded in the cast walls thereof wrought iron reinforcing members 8, 10 and 12, for the purpose of preventing cracking of the box in the repeated heating and cooling to which the box is subjected. These wrought iron reinforcements may be variously distributed at points of known weakness in the box and preferably are located so as to reinforce the corners of the box. In the production of these boxes, the mold is formed in the usual way and the

wrought iron inserts, such as 8, 10 and 12, are positioned in the mold by means of chaplets, and the metal to form the walls of the box is heated



to a point sufficiently high to fuse the wrought iron surfaces when poured around them, so as to form a weld with the surface of the wrought iron members when the casting is poured.

1,582,353, Magnetic Alloy, Willoughby Statham Smith, of Benchams, Newton Poppleford, and Henry Joseph Garnett, of Seven Oaks, England.

This patent describes a new magnetic alloy which has a low hysteresis loss with a high permeability at low magnetizing forces, such as the magnetizing forces present in telegraph and telephone cables. The alloy comprises nickel at least 71 per cent of the whole, iron between 19 and 21 per cent and copper between 5 and 6 per cent. To increase the electrical resistance, a fourth element, such as tungsten, chromium, silicon, vanadium, titanium, molybdenum or aluminum may be added preferably in amounts of less than 1 per cent. An alloy comprising nickel 73.0 per cent, copper 5.4 per cent, iron 20.7 per cent, tungsten 0.6 per cent and manganese 0.3 per cent has an electrical resistance of 25 microhms per cubic centimetre and an initial magnetic permeability of 6600.

1,586,543, Process of Refining Metals; Sigurd Westberg, of Germantown, Pennsylvania.

This patent describes a process of reducing or removing certain impurities from metals and alloys while in a solid state, and more particularly the removing of sulphur and carbon from metal rods, sheets, bars, or the like, by coating the metal with lime or an oxide of an alkaline earth, stacking the same in a heating chamber or muffle, passing hydrogen gas into the chamber until all of the oxidizing atmosphere is removed and then heating to a sufficient temperature to cause a combination of the hydrogen with the impurities in the metal. The temperature of this combination will be between 800 and 1400 degrees Cent. The pressure on the gas in the muffle may be varied to vary the rate of removal of the sulphur and carbon.

1,586,590, Manufacture of Ferrochromium Alloy; 1,586,591, Manufacture of Ferro-Alloys, Particularly Ferrochromium Alloys; 1,586,592, Manufacture of Unstainable Irons and Steels, Ronald Wild, of Holmesfield, England, Assignor, by Mesne Assignments, to Rustless Iron Corporation of America, a corporation of Maryland.

This group of patents relates to the production of ferrochromium al-

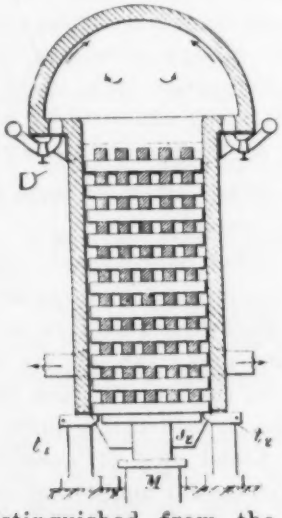
loys and to the use of these alloys in the manufacture of rustless or unstainable iron or steel. According to the first patent, a ferrochromium, containing a low percentage of carbon and silicon suitable for producing rustless iron as well as rustless steel, is produced by subjecting a base metal, comprising iron and chromium and containing impurities, such as carbon and silicon, to an oxidizing blow in order to reduce the carbon and silicon to the desired point. The novelty of the process comprises in adding to the base metal an additional oxidizable element, such as manganese, which will be largely removed in the oxidizing blow, but which will protect the chromium of the alloy from oxidation during the blowing operation.

The second patent describes a method of producing ferro alloys, such as ferrochromium, with a low carbon content by the reduction of the oxides of the metals through an exothermic reaction of the silicon, the novelty of which resides in carrying out the reaction with an excess of the oxide of the metal to be reduced over the chemical equivalent of the amount of ferrosilicon used.

The third patent describes the production of an unstainable iron or steel by the mixture of molten crude ferrochromium with molten iron or steel of approximate purity in proportions to provide an initial product of substantially the composition of the desired unstainable metal and then purifying this product by means of an oxidizing blow which removes the silicon contained in the crude ferrochromium.

1,586,599, Vertical Heating Furnace, Ernst Bohler, of Rosselange, France.

This patent describes a vertical heating furnace in which the materials to be heated are moved counter to the current of heated gases in



a vertical chamber as distinguished from the same type of continuous horizontal furnace, thereby securing a more uniform heating of the article than is possible in the horizontal furnaces of this type, because in

(Continued on Page 139)

The Question Box

A Column Devoted to the Asking, Answering and Discussing of Practical Questions in Heat Treatment — Members Submitting Answers and Discussions Are Requested To Refer to Serial Numbers of Questions

QUESTION NO. 190. *What is the most suitable thickness of a copper plate coating to prevent carburization in the box type method of carburizing.*

QUESTION NO. 189. *What is the McQuaid-Ehn test and what are its merits in selecting steel for carburizing?*

QUESTION NO. 188. *What causes the scaling on the inside of holes of piston pins when heated in a Hoskins electric pusher type furnace? Does the sulphur content of the steel have any marked effect?*

QUESTION NO. 187. *What causes the carbonaceous deposit on steel parts when quenched in oil from temperatures about 1600 degrees Fahr.? How can this be eliminated? Do all oils have the same effect in the same degree?*

QUESTION NO. 186. *Why should marking dies be rubbed with emery cloth before using?*

QUESTION NO. 185. *What precaution should be observed in designing such a tool?*

QUESTION NO. 184. *What precaution should be observed in quenching tool steels containing chromium?*

ANSWER. By S. P. Rockwell, Hartford, Conn. The main precaution to be used in quenching tool steels containing chromium is to be absolutely sure that full saturation of heat has been secured and the work has completely and uniformly emerged from the critical range. The Rockwell dilatometer shows steels containing enough chromium to demand an oil quench to have entirely different characteristics than those steels which can be safely quenched in water. This is chiefly observed by an entire lack of contraction or expansion after the main contraction of the critical range has been passed.

QUESTION NO. 183. *What effect, if any, have the additions of alloys on the internal stress of tool steels?*

QUESTION NO. 182. *What indication, if any, is given in the quenching of tool steel if the steel is quenched while going through the critical range?*

ANSWER. By S. P. Rockwell, Hartford, Conn. In working with the Rockwell dilatometer any steel going through its critical range contracts, and if this steel is of such a nature that it will crack in the quenching, if quenched during this contraction period, it will in the great majority of cases break in the quench. It is usually found that if quenched during this contraction period the steel quenched will squeal more than ordinary.

QUESTION NO. 181. *How may a pyrometer be checked by means of carbon tool steel?*

ANSWER. By S. P. Rockwell, Hartford, Conn. In any straight carbon tool steel it has been found that the contraction period by the Rockwell dilatom-

eter indicates the Ac_1 critical temperature, and if a pyrometer is used in conjunction with the heating, it will be found that the contraction takes place at a temperature of 1355 to 1360 degrees Fahr. Of course, the pyrometer fire end must be in contact with the work. On slow heating this is found to be quite a reliable check.

QUESTION NO. 180. *Explain the effect of decarburization in grinding.*

QUESTION NO. 179 *What is the difference between grinding cracks and grinding checks?*

QUESTION NO. 175. *How long should high speed steel tools be left in the high draw?*

ANSWER. By S. P. Rockwell, Hartford, Conn. The answer given by E. P. Geary, to leave the steel in one hour after saturation of the drawing heat has been obtained is good practice. To determine when the saturation period has been completed is readily done by the Rockwell dilatometer. When no contraction or expansion is shown by the indicating dial or recorder, then there can be no interchange of heat between the furnace and the work being tempered, and in consequence thorough saturation has been secured.

QUESTION NO. 174. *How long should high speed steels stay in the preheat?*

ANSWER. By S. P. Rockwell, Hartford, Conn. High speed steel should stay in the preheat until full saturation can be secured. This can be readily done by the Rockwell dilatometer as described in answer to Question No. 175.

QUESTION NO. 171. *What is the hardness of carbon steels at elevated temperatures such as 400 to 1000 degrees?*

QUESTION NO. 170. *Why do you get a low Rockwell reading on Stellite?*

ANSWER. By S. P. Rockwell, Hartford, Conn. The main reason for a low reading of Stellite is undoubtedly due to the crystallization and dendritic structure obtaining in this metal. The metal seems to crumble under the Brale test and form a rough crater entirely opposite to that formed by standard steels.

QUESTION NO. 169. *What is the hardness of high speed steel at elevated temperatures such as 400 to 1000 degrees?*

QUESTION NO. 165. *What importance should be placed on the carbon content of high speed steel?*

QUESTION NO. 159. *What is the explanation for the retarding effect on corrosion of small amounts of copper in low carbon steel?*

ANSWER. By C. G. Williams, metallurgist Barber Colman Co., Moline, Ill.

What seem to be the facts and what is put forward as a tenative solution of the proposition concerning the resistance of copper bearing steels to corrosion, will be given as the true solution, until some other better solution is put forward.

The steels tested have been those low in carbon, that is, having from 0.10 to 0.20 per cent carbon, from 0.005 to 0.015 per cent sulphur and with 0.50 to 0.65 per cent copper.

In these types of steels, some of the copper units with part or all of the sulphur to form copper sulphide, Cu_2S, and the remainder of the copper forms a solid solution in the ferrite. Micrometallurgical examination shows that this copper sulphide is at the boundries of the grains and entirely surrounding them. The crystals or grains have an irregular structure with curved outlines. This condition seems to have been brought about by the grains in the state of formation being covered with a coating of the copper sulphide. The grains are thus caused to assume a finer structure and the pearlite will so closely approach to the martensitic form as to be almost undistinguishable from it.

Because of the fact that the small ferrite crystals are completely surrounded by the coating of copper sulphide, which is almost impervious to moisture or to the action of certain corroding gases or their combined action, the moisture in the air or in combination with those gases cannot come in contact with the molecules of iron to cause them to oxidize, and therefore corrosion of the metal can be carried on to only a very slight extent and this only where the crystals or their surrounding envelope of copper sulphide has been broken or crushed as in rolling or other forming work or entirely removed over a portion of the molecule as in mechanical operations of other types.

Copper bearing steels are not properly corrosion resistant, for they will corrode to some slight extent, as noted above, but in proportion to the amount of corrosion exhibited by plain carbon steels and most types of alloy steels, it is so slightly affected as to appear to be entirely corrosion resisting.

QUESTION NO. 158. *What is the present recognized cause for the "transverse fissure" as occasionally found in rails?*

QUESTION NO. 157. *Does 0.070 to 0.090 per cent phosphorus in low carbon basic open hearth steel tend to prevent "stickers" in the rolling of sheets? If so, what is the explanation for such a tendency?*

Abstracts of Techincal Articles

(Continued from Page 131)

IRON. By Tario Kikuta: Kinzoku no Kenku. Vol. 3, No. 4, 1926, p. 185-222.

The author investigated the details of the change of white cast iron to black heart malleable cast iron by means of a dilatometer.

Since the graphitization of a cast iron is always accompanied by a considerable expansion of its volume, the progress of graphitization at high temperatures can be followed continuously by measuring the expansion of the cast iron at these temperatures. He classified the graphitization into the first and second stages, that is,

(1) the graphitization of free cementite above the A_1 point, that is, of primary (if present) and eutectic cementites;

(2) the graphitization of eutectoid cementite precipitated from austenite at the A_1 point,

and investigated the relation of the graphitization to temperature or to heating interval. The results are as follows:

1. To obtain a black heart malleable cast iron from a white cast iron, it is necessary first to heat the sample for a prolonged time at a high temperature above the A_1 point, and then to heat it for a proper time

just below the A_1 point to graphitize the eutectoid cementite separated out at this point.

2. The time intervals required for the first and second stages of graphitization depend considerably upon temperature. The relation between the heating hours (t) and the temperature T (T in absolute scale) is given by:

first stage of graphitization: $t = 6.87 \times 10^{9}C^{-0.0161\ t}$

second stage of graphitization: $t = 5.11 \times 10^{5}C^{-0.0117\ t}$;

for a sample containing 2.56-2.74 per cent carbon and 1.12-1.29 per cent silicon.

3. The size of the cementite particles in cast iron depends on the thickness of the cast piece, or more correctly, on the velocity of cooling after the casting; the thicker the cast, the larger are the cementite particles, and hence the graphitization is more difficult.

4. The pouring temperature of white cast iron affects its graphitization; the higher this temperature, the more difficult is the graphitization.

5. The author also investigated the effect of a small quantity of a third element on the graphitization. Silicon favors the first and second stages of graphitization. Carbon affects a little the first stage of graphitization; but considerably favors the second stage. Manganese resists the graphitization; the effect is not so large for the first stage of graphitization, but considerable for the second stage. The same remark applies also for the case of sulphur. Phosphorus a little favors the first stage of graphitization; but for the second stage its quantity must be smaller than a certain value (0.3 per cent phosphorus.)

Abstracted by Dr. Kotaro Honda.

Reviews of Recent Patents

(Continued from Page 135)

the horizontal furnaces the heated gases concentrate at the top of the chamber, and only the upper portions of the billets or the like are heated. The apparatus comprises a vertical shaft provided at the top thereof with a dome-shaped heating chamber into which the heated gases are forced by suitable pipes. At the bottom of the vertical shaft, an elevating ram or pusher M, which may be hydraulically operated, is used to move the stacks of billets progressively upward. The billets are laid in spaced relation upon horizontal supports and pushed upwardly by the ram M into the base of the stack where they pass the supports T-1 and T-2, which permit the passage of the billets supports upwardly, but will not permit the downward movement of the billet supports from the stack. The ram M is then withdrawn, a fresh layer of billets placed thereon and pushed into the base of the stack, thereby elevating the first layer off the supports T-1 and T-2, and upon withdrawal of the ram M depositing the second layer of billets upon these supports, and so on, each operation of the ram M elevating the stack of billets toward the top of the vertical chamber in counter current to the heated gases which enter at the top of the chamber and escape from the flues at the base. When the stack of billets in the chamber reaches the height of the door D near the top, this door is opened and at each operation of the ram M, one layer of the heated billets removed.

News of the Chapters

STANDING OF THE CHAPTERS

IN the June issue of TRANSACTIONS appeared the relative membership standing of the 29 chapters of the Society as of April 1, 1926, and as of May 1, 1926. The tabulation which appears below shows the relative membership standing of the chapters on May 1, 1926, and June 1, 1926.

STANDING OF THE CHAPTERS AS OF JUNE 1, 1926

GROUP I	GROUP II	GROUP III
1. Detroit .....(386)	1. Hartford .....(138)	1. Los Angeles ...(74)
2. Cleveland .....(376)	2. Lehigh Valley..(137)	2. Tri City .....(66)
3. Philadelphia ..(356)	3. Golden Gate...(129)	3. Rockford .....(62)
4. Pittsburgh ....(317)	4. Syracuse .....(101)	4. Washington ...(61)
5. Chicago .....(312)	5. Milwaukee ....(100)	5. Worcester .....(57)
6. New York ....(255)	6. Cincinnati .....(83)	6. New Haven.....(56)
7. Boston .....(234)	7. Indianapolis ...(82)	7. Rochester .....(55)
	8. St. Louis.....(76)	8. Schenectady ...(47)
	9. Buffalo .....(56)	9. Toronto .....(46)
	10. North-west ....(48)	10. Rhode Island ..(43)
		11. Montreal .....(35)
		12. Fort Wayne....(32)
		13. Springfield ....(30)

STANDING OF THE CHAPTERS AS OF MAY 1, 1926

GROUP I	GROUP II	GROUP III
1. Detroit .....(384)	1. Lehigh Valley.(133)	1. Tri City .....(64)
2. Cleveland ....(375)	2. Hartford .....(132)	2. Los Angeles ...(64)
3. Philadelphia ..(353)	3. Golden Gate...(132)	3. Rockford .....(59)
4. Pittsburgh ...(321)	4. Syracuse .....(100)	4. New Haven ...(58)
5. Chicago .....(306)	5. Milwaukee ....(100)	5. Washington ....(58)
6. New York ...(251)	6. Cincinnati .....(80)	6. Worcester .....(57)
7. Boston .....(236)	7. St. Louis .....(75)	7. Rochester .....(51)
	8. Indianapolis ...(75)	8. Schenectady ...(49)
	9. Buffalo .....(56)	9. Toronto .....(45)
	10. North-west ....(50)	10. Rhode Island ..(43)
		11. Springfield ....(30)
		12. Fort Wayne ..(26)

THERE are now over 4000 members of the A. S. S. T., the actual count on the first of the month being 4011, a net gain of 83 new members last month.

GROUP No. 1—All but one of the chapters in this group showed a gain,—

the largest being Chicago with a net increase of 6.

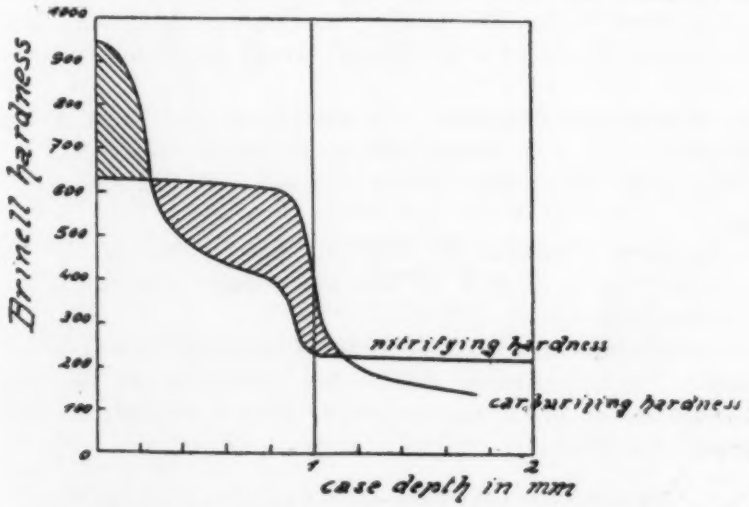
GROUP No. 2—Lehigh Valley had the honor of holding first place in Group 2 for a month; however, there is a long span of future ahead and, judging from Lehigh Valley's activities, they will not be content to let Hartford occupy the coveted position. The standing of the other chapters in the group remains the same, with the exception of Indianapolis with an increase of 7 new members, passing St. Louis.

GROUP No. 3—Greetings to Los Angeles—the new leader in this group. For a great many months Tri City has headed this list with Los Angeles either tied with them, or just a few members behind. However, Los Angeles has turned over their membership work to a very active membership committee and they had an increase of 11 new members last month, which gives them now a lead of 8 over Tri City. Rockford gained three new members after their remarkable advance last month, which now gives them a membership of 62. New Haven dropped from fourth position to sixth, which permitted Washington, with a gain of 4, and Worcester with no gain, to occupy positions 4 and 5.

It will be observed that a new group has been admitted to GROUP 3, with 35 new members,—that is Montreal,—Fort Wayne, the other new group, having been admitted last month. Montreal had 24 completed memberships during the past period and Fort Wayne had an increase of 6.

BOSTON CHAPTER

At the annual meeting of the Boston Chapter, May 7, 1926, Dr. G. Waterhouse, professor of metallurgy at Massachusetts Institute of Technology, who had recently returned from a business trip in Germany, briefly described the Krupp Nitride Case-hardening Process, describing what he saw on his visit to Essen and quoting from Dr. Fry's article in the February, 1926, issue



of the Krupp Monatshefte. Special steels having the same mechanical and machining properties as our nickel and nickel-chromium steels have been developed by Krupp especially for this purpose and are the only ones suit-

able for the process. The parts are finish machined and in cases where it is thought necessary, they are annealed to remove any strains. They are then acted upon by gases, chiefly ammonia, which liberate the nascent nitrogen at moderate temperatures. In no case does the temperature exceed 1076 degrees Fahr. After the necessary length of time to give the desired depth of penetration, the parts are slowly cooled and are ready for use without further heat treatment.

The surface hardness is very high, experiments indicating that it approaches 950 Brinell against about 630 for regular hardened carburized surfaces. This hardness, however, falls off rather quickly so that at a depth of 0.2 millimeters, the carburized parts are much harder than those nitrified. For this reason the process is not suitable for parts exposed to surface pressures of over 100,000 pounds per square inch. A particular feature of the process is the fact that fragile, delicate pieces can be treated with no deformation if they are free from internal strains before treatment.

Many examples of parts which were being commercially treated by nitrifying were given, including automobile and typewriter parts, bolts, gages, and so forth. When asked how such high Brinell figures were determined, Dr. Waterhouse said that the actual tests were made with the Herbert Hardness Tester and the Brinell figures calculated from the results thus obtained.

This is the second coffee talk which Dr. Waterhouse has given the chapter since his recent trip to Germany and the members are looking forward to hearing more of his interesting experiences there at a future meeting.

H. E. Handy.

CINCINNATI CHAPTER

The regular May meeting of the Cincinnati Chapter, A. S. S. T., was held at the Ohio Mechanics Institute on the 12th day of May. There were 45 in attendance.

The chairman appointed C. M. Bigger, chairman, A. J. Lucas, E. F. Ries as members of the "On To Chicago" committee, whose duty it shall be to stimulate interest in the national convention and exposition to be held in September.

The "Educational Committee," consisting of Prof. G. M. Enos, chairman, W. R. Klinkicht, and A. J. Lucas, was also appointed, whose duty it shall be to arrange details in connection with the proposed course in metallurgy and heat treatment.

The secretary submitted the following report: During the year 1925-26 the Cincinnati Chapter, A. S. S. T., held nine regular meetings and four executive committee meetings.

The average attendance at the monthly meetings was about 45 members and guests. The approximate membership attendance at the meetings was about 35 members or 33-1/3 per cent. The total attendance for the year was 403 members and guests.

MEETINGS—DATES—SPEAKERS—SUBJECTS

Sept. 10, 1925—Geo. Desautels, Heppenstall Forge & Knife Co. "The Manufacture and Heat Treatment of Die Blocks," illustrated with a motion picture. Held at the University of Cincinnati.

Oct. 8, 1925—G. C. Davis, New England sales manager, The Bellis Heat Treating Co. "The Application of Lavite to Steel and Non-Ferrous Heat-Treatment," illustrated with slides. Held at the Ohio Mechanics Institute.

Nov. 12, 1925—Sam Tour, Doehler Die Castings Co. "Die Castings," illustrated with slides and samples. Held at the Ohio Mechanics Institute.

Dec. 18, 1925—J. M. Watson, metallurgical engineer, The Hupp Motor Car Co. "The Heat Treatment of Automobile Forgings," illustrated with a motion picture. Held at the University of Cincinnati.

Jan. 14, 1926—Mr. Strickland, The American Rolling Mill Co. "The Corrosion of Ferrous Metals." Held at the Ohio Mechanics Institute.

Feb. 11, 1926—J. Korp, The Leeds & Northrup Co. "The Hump Method of Heat Treatment." Held at the Ohio Mechanics Institute.

Mar. 11, 1926—Prof. G. M. Enos, University of Cincinnati. "The Spark Testing of Steels," illustrated with slides and a practical demonstration. Held at the Ohio Mechanics Institute.

April 15, 1926—W. R. Fleming, The Andrews Steel Co., metallurgical engineer. "The Relation Between the Steel Maker and the Steel Treater." Held at the Ohio Mechanics Institute.

May 12, 1926—J. Cran, Supervisor of Heat Treatment, The Peters Cartridge Co. "Why Steels Warp and Crack." Held at the Ohio Mechanics Institute.

The following were duly elected officers for the ensuing year:

Chairman, W. R. Klinkicht, The Pollak Steel Co.

Vice-Chairman, A. J. Lucas, The Cincinnati Gear Co.

Secretary-Treasurer, Fred L. Martin, Nat'l Metal Trades Ass'n.

Executive Committee: E. P. Stenger, The Queen City Steel Treating Co.; N. M. Salkover, The Queen City Steel Treating Co.; Prof. G. M. Enos, University of Cincinnati; H. S. Binns, The Cinti Milling Mach. Co.; J. H. Nead, The American Rolling Mill Co. (Middletown, O.)

J. Cran, supervisor of heat treatment, The Peters Cartridge Co., was then introduced and presented a most interesting paper entitled "Why Steels Warp and Crack." His subject was selected from an article under the above heading appearing on page 373 of the March issue of TRANSACTIONS.

Mr. Cran's many years of practical experience devoted to the solution of shop problems in heat treatment enabled him to cite many difficult problems and the proper method of overcoming failure of tools, dies and other parts that fail through cracking and distortion.

The paper was followed by a lively discussion which terminated only on account of the lateness of the hour. *Fred L. Martin.*

CLEVELAND CHAPTER

The May meeting of Cleveland Chapter was held May 21, 1926, at 8:00 p. m., in the Engineering Society Rooms at Hotel Winton. The program for the evening was the showing of a six reel motion picture taken at the Youngs-

town Sheet and Tube Company, Youngstown, Ohio, and presented by G. A. Reinhardt, metallurgist for that company.

This picture showed in detail the manufacture of pig iron, bessemer steel and basic open hearth steel and some methods of manufacturing pipe. The film was both entertaining and instructive, and, although it was made under necessarily difficult conditions, the photography was excellent.

This was the last meeting of the season for Cleveland Chapter. Reports were made by the retiring chairman and secretary-treasurer. The election of officers and members of the executive committee resulted as follows:

Chairman—W. P. Sykes, National Lamp Works

Vice-Chairman—A. S. Townsend, Cleveland Twist Drill Co.

Secretary-Treasurer—A. E. Buelow, Lamson & Sessions Co.

Executive Committee—James Ayling, Case Hardening Service Co.; G. C. Beebe, Columbia Tool Steel Co.; E. W. Goodaire, Cleveland Frog & Crossing Co.; H. A. Schwartz, National Malleable & Steel Castings Co.; V. L. Tichy, White Motor Co.; W. H. White, Atlas Steel Corp. *A. S. Townsend.*

DETROIT CHAPTER

The April meeting of the Detroit Chapter of the American Society for Steel Treating was held Friday, April 9, 1926, and was a joint meeting with the Detroit Engineering Society. The usual Coffee Talk was given by two speakers. The first speaker was G. C. Dilleman of the State Highway Department, Lansing, Michigan.

Mr. Dilleman said that the Highway Department started in 1905, when a 19-foot road was considered ideal. In 1913 state trunk lines were inaugurated, but up to 1919 county-made roads were the rule. Michigan has 77,000 miles of highway, 69,000 under the State Highway Department, 21,000 miles of improved roads, of which 4000 miles are paved. Michigan has 5200 miles of Federal Aid Highway, and it will not be long before people will think in terms of federal roads only.

The new construction roads, i. e., those less than 5 years old (all 5 to 10-year-old roads are out of date), will stand truck traffic up to 14 tons, maximum load. These roads are thicker, and 20 feet minimum, most roads and all bridges are 30 to 40 feet wide. Then there is starting the super highways, which are 204 feet wide. One mile of 20-foot roadway (concrete) costs $35,000, but gravel roads cost 2½ times as much upkeep, so that concrete is cheaper. Maintenance of present roads costs the State of Michigan $4,000,000 a year. The present program calls for decreasing accidents by grading curves, by grade crossing eliminations and proper signs along the highway.

The second speaker, George Appel, spoke on the Boy Scout Movement. Mr. Appel gave a good idea of the army of boys of high order living by scout laws. The Boy Scout program is an educational system. Boys need to be bound together to strive for something better and on this idea and a boy's instinct for play the Boy Scout movement has grown so much that several colleges, as Columbia, Wisconsin, California, etc., have special courses for Scout Masters. The Boy Scout system runs on the merit system and the boy must learn to earn his insignia. The movement

is 16 years old, of English origin, and the results are now being felt. Over 40 per cent of the freshmen of Annapolis and Harvard were Boy Scouts. The scouts are making good and have their foundation in manhood.

The speaker of the evening was Dr. Zay Jeffries, our National Treasurer, and head of the Research Bureau of Aluminum Company of America. Dr. Jeffries spoke on "Engineering and Science in the Metal Industry."

Dr. Jeffries started his paper by giving a brief perspective of the metal industry. Iron, it seems, constitutes 93 per cent of the total metal production. Other metals of the 50 chemical elements used in industrial alloys to make 99.5 per cent of the world's production are copper, lead, zinc, aluminum, tin, and nickel. The value of metals produced in small quantities to humanity as silver, chromium, tungsten and plantinum, was brought out. We could afford to pay $330,000 a pound for tungsten rather than go back to carbon filament lamps.

Dr. Jeffries then traced the growth of the metal industry since 1885. The gradual increase in the use of nonferrous material was in evidence. In the steel industry the increasing use of alloy steel is most evidenced. As a result of these two facts we find a pound of metal is being made to give more service than ever before.

The problems of selecting the best metal or alloy for a given purpose is the great question of the engineer. This is a complex problem. In many cases there is only one base metal deserving consideration because of cost, quantity, shapes and sizes. No engineer is needed to determine that steel is the proper material for rails. Engineering experience is required to select the proper kind of steel and the proper treatment so the rail will have a suitable life. Assuming that several materials will function the selection should be made on the basis of ultimate economy. This is a selection by economic compromise and is not always obvious.

Dr. Jeffries stated that up to 50 years ago the metal industry was all art and no science. It is because of science that the metal industry has progressed more in the last 50 years than the previous 10,000 years. In these days it does not take long to correlate and interpret observations. As an illustration of this, Dr. Jeffries gave the following:—Pure iron is heated in hydrogen to 1450-1500 degrees Cent. and quenched in oil and becomes glass hard. Quenched in water, it is soft, or heated in air and quenched in oil, it is soft. In less than an hour the microscope explains all. The iron has a white iron structure on the surface and normal carbonless iron on the inside. The explanation is easy and immediate. Hydrogen keeps iron clean, when thin clean iron strikes the oil some of the carbon combines with the iron and lowers its melting point. Before the iron can cool, low enough carbon is absorbed on the thin melted surface layer and the iron carbon alloy is hard.

The classification of metals and alloys by Dr. Jeffries was:

1. Pure Metal
2. Solid solutions
3. Compounds

An alloy, no matter how complex, is made up of one of these and it is a metallurgical problem to recognize items. A metallurgist should also determine the properties of the constituents of an alloy separately, because only

by so doing is he able to tell the effect of each constituent on the properties of the alloy.

The factors governing the properties of an alloy will depend on:

1. The number and properties of the constituents present
2. The proportions of the constituents
3. The arrangement of the constituents

These properties can be further explained on atoms and variations in their arrangement. The sum of the atomic attractions on a plane through a piece of metal normal to the stress may be called absolute cohesion of the metal. The tensile strength found is always lower than the absolute cohesion. In fact, softness and weakness are found in spite of high cohesion values. The key to their interpretation is found in crystal structure. Permanent deformation of a solid metal is known to take place largely by slippage of one portion of a crystal with reference to another along crystallographic planes. There are a number of ways of increasing slip interference; in a solid solution, or pure metal by grain refinement and by hard metallic compounds in other alloys.

By means of slides Dr. Jeffries gave a schematic representation of a section through a crystal of solid solutions and one of a crystal in which a portion of the solute material has combined and formed a hard metallic compound scattered through the area. He then showed the effect when all the metallic compound grew into one large portion. In the latter case, slip could take place almost as easy as in the solid solution and much easier than when the compound was scattered through the crystal.

Dr. Jeffries then applied this reasoning to the hardening of steel. By use of X-rays it has been proven that martensite is very fine grained. It is therefore evident that in this case we have hardness and reluctance to slip due to

1. Ultra-fine grain size
2. Carbon in solid solution

The maximum hardness of any alloy can be obtained by super grain refinement or critical dispersion of the hard metallic compound. In steel the grain refinement is sufficient to almost account for martensite hardness and the carbide precipitation has little effect. By a reasonable interpretation of these facts the hardening of high speed steel was explained.

The next point to which Dr. Jeffries called attention was a 22 per cent molybdenum iron alloy. On slow cooling an iron molybdenum compound precipitated out, if quenched from 1400 degrees Cent., a solid solution results, and if tempered at 600 to 650 degrees Cent., a hard alloy results which appeared trosto-sorbitic on the slide shown. The material was coarse-grained but the 600 degrees Cent. draw gave critical dispersion and the material, from 214 Brinell as quenched, became 531. Dies from this alloy run at a red heat lasted 60 times as long as high speed steel. It is significant that this heat treatment is typical of the nonferrous rather than ferrous alloys.

In closing Dr. Jeffries said that as time goes on we will see an extended use of alloy and heat treated steel, that consumers are being supplied with a better product every day and less new metal is required for a given service to the consumer than at any time in history. In harmony with this idea is

mass production and standardization. We look for mass production in the metal industry, but it must and will be mass production of quality products.

A rising vote of thanks was tendered Dr. Jeffries for his splendid paper and forceful presentation. ——— *F. P. Zimmerli.*

A precedent was established for the American Society for Steel Treating, by the Detroit Chapter in initiating extension meetings. The need for holding society meetings at Ann Arbor was sensed by Dr. A. E. White and other Ann Arbor members who put the matter before the Detroit Executive Committee. After some consideration the idea was brought forth by "Bob" Atkinson, that the Detroit Chapter might sponsor extension meetings at Ann Arbor and other places. These meetings should not interfere with the regular Detroit schedule of meetings, but should serve and stimulate membership in the outlying districts. After some discussion this idea was accepted and a series of four meetings for the year planned for Ann Arbor. The first of these meetings was held April 29th and proved that there is a great deal of interest in the vicinity of Ann Arbor. Over 100 men turned out for this meeting, of which about 15 were from Detroit, the remainder being composed of men from Jackson, Lansing and Ann Arbor.

Dr. A. E. White opened the meeting with a short introductory talk. For the benefit of the large number present who were not members, he outlined the history of the society and the service rendered by it.

The principal speaker of this meeting was R. S. Archer of the Aluminum Company of America, who spoke on "Recent Developments in the Hardening of Metals."

Mr. Archer gave a very concise and clear explanation of the slip interference theory of hardening. According to Mr. Archer, relative hardness is caused by particles which interfere with the potential planes of slip of the crystals of alloys. There is a critical size of particle which produces a maximum hardness for a given alloy by offering interference to the maximum number of planes of slip. If this alloy is heated these particles coalesce, resulting in larger and fewer particles and corresponding decrease in interference to the slip in the crystals, and consequently decrease in hardness. Mr. Archer used the theory to account for the hardening of aluminum alloys by quenching and by subsequent aging. He further applied his theory to explain the hardness of steel, as well as the alloys of copper-iron, copper-zinc, lead-antimony, iron-tungsten, and iron-molybdenum. The latter two alloys are examples of hardening iron without the introduction of carbon and in which allotropy plays no part.

After the meeting lunch was served to give an opportunity for discussion and visiting. ——— *W. E. Jominy.*

The annual spring meeting of the Detroit Chapter of the American Society for Steel Treating was held at the Michigan Union in Ann Arbor, Michigan, on May 22, 1926. After the announcement of the names of the newly elected officers of the chapter for next year and the treasurer's report, Chairman Calkins turned the meeting over to Dr. A. E. White, who introduced the speaker of the evening, C. F. Kettering of the General Motors Company. Mr. Kettering's paper will be published in a later issue of TRANSACTIONS.

GOLDEN GATE CHAPTER

The following is a report of the Golden Gate Chapter meeting, held on February 10, which has been delayed in publication, due to the fact that the original copy of the report was lost in the mails. A copy of the original report was supplied by the secretary, G. Hanson Grubb.

The February meeting of the Golden Gate Chapter, held Wednesday evening, February 10, was a most unqualified success. It was in the form of a trip to the Pittsburg plant of the Columbia Steel Corporation, which is located some forty miles from San Francisco. We had a special train of three cars and took up one hundred thirty-seven members and guests.

On the train the party was divided into groups of five, each individual being tagged and notified that he was one of a small party to be in charge of a guide during the evening. On arrival at Pittsburg, we were met by the executive of the plant and taken at once to the Hotel Los Medanos for dinner. A number of other members who had driven up in automobiles joined us at this point and dinner was immediately served.

Our chairman, F. B. Drake, opened the meeting at once, with an explanation of the purpose of the trip and tribute to the courtesy of the Columbia Steel Corporation in making possible this plant visit, and introduced N. A. Becker, general superintendent of the plant. Mr. Becker responded for the Columbia organization, welcoming the visitors and giving a short history of the Columbia organization and development. He then introduced the various officers of the plant, who were to act as guides during the evening, and presented his assistant superintendent, Mr. Cohn, who was directly in charge of the arrangements.

For the benefit of strangers and guests of members, who were present, Mr. Drake gave in a few words, the history of the organization since the formation of Golden Gate chapter, and described the educational work that is in progress, both through monthly meetings and through our course in ferrous metallurgy. He concluded by introducing the various officers of the chapter who were present. Among these were A. N. Armitage, metallurgist of the Columbia Steel Corporation, and member of the executive committee of the chapter, to whose initiative the chapter owes its organization, and who had devoted much time to the arrangements for this trip. He also introduced Fred Wight, another member of the executive committee, of the U. S. Steel Products Corporation, as in general charge of the trip on the part of the chapter. Other members of the executive committee and officers present were also introduced to the gathering.

Immediately after dinner, the visitors were taken by busses to the plant and segregated into the previously arranged groups of five, each in charge of an official. With those who had driven directly to Pittsburg by automobile, the total number of visitors, entirely excluding Columbia members, was one hundred sixty, a record crowd for any event since the organization of Golden Gate chapter.

The parties were first taken to the open hearth department. The steel made here is from scrap and pig iron produced by the company's own blast

furnace in Utah, where they also have iron mines, coal mines, and coke ovens.

The open hearth plant consists of two 60-ton basic, two 20-ton basic, and one 7-ton acid furnace, producing an average of 9,000 tons per month.

We then inspected the foundry, which is believed to be the largest steel foundry west of the Rocky Mountains, and produces mild and alloy basic open hearth castings from 2 to 80,000 pounds each, with a capacity of a thousand tons per month.

Passing through the power plant and the first aid hospital, and through the machine shop, we arrived at the rolling mill department, and first inspected the 18-inch roughing mill, which has a capacity of 9,300 tons per month. This produces billets for the merchant and rod mills and sheet bar for the sheet mill. Then past the 12-inch merchant mill to the rod mill, where they were rolling a 2¼ x 2¼-inch billet into No. 5 rod. The capacity of this mill is 3,300 tons per month. The lights were turned out to show the dramatic effect of the red hot rod moving through the darkness.

The wire mill department was next, where we saw a number of different kinds of wire being cold drawn. In addition to the ordinary kinds of wire, this mill produces galvanized, barbed, and copper, and has a capacity of 3,000 tons per month.

The nail mill was next on our route, operating 54 nail machines, producing sizes from 2-penny to 7-inch spikes.

The final unit was the sheet mill, which produces the various types of sheet, and was the first sheet mill to operate west of the Mississippi River. It handles sizes ranging from 14 to 30 gage, with a capacity of 2,500 tons per month. A box was drawn from one of the annealing furnaces for the benefit of the visitors.

This Pittsburg plant employs approximately 1150 men. Including the coal mine, iron mines, coke ovens, blast furnace in Utah, and the open hearth and electric furnaces, steel foundry and rolling mill in Los Angeles County, and the electric furnace and steel foundry in Portland, Oregon, the organization employs a total of 2,500 men.

By this time we had been in the plant over two and a half hours and had visited every important section, although, of course, a much greater length of time for inspection would have been appreciated by the members. Our train was waiting for us, and we left Pittsburg at 10:55 p. m., arriving in San Francisco shortly after midnight.

This has been a noteworthy occasion in the history of Golden Gate chapter, not only on account of the largest crowd that we have ever handled, but from the great interest displayed by the members and their friends in the privilege of inspecting the largest steel plant on the Pacific coast. *D. Hanson Grubb.*

The fifteen weeks' Course on Ferrous Metallurgy which was sponsored by Golden Gate Chapter has just been completed with an extremely good attendance and with universal expressions of satisfaction on the part of the members of the class. The committee in charge, consisting of C. S. Moody, chairman, H. S. Taylor, and E. H. Kottnauer, have given very generously of their time and effort and the result has surpassed the most sanguine expecta-

tions. The course proper ended with the session of Monday evening, April 26, but a supplementary meeting was held May 3, for the purpose of accomplishing several desired results. In the first place a question box was held and a number of puzzling points cleared up to the satisfaction of the students. Secondly, a written quiz arranged in the form for a positive or negative answer and covering the whole course, was given . Third, an unsigned expression of opinion on the best and least attractive points of the course, and suggestions as to future work of the same kind. This last query showed a remarkable unanimity of opinion in praising the work of the instructors.

Some of the members of this class came regularly to meetings in Oakland or San Francisco from plants located forty miles away from the class room or laboratory, as the case might be, and the members of the class were drawn not only from San Francisco and Oakland, but from Berkeley, Richmond, Pittsburgh, San Leandro, Mare Island Navy Yard, and last, but not least, San Carlos.

At the end of the evening, the class took charge and informed the instructors that in view of its desire to show its appreciation of the work that had been done, it wished to have these gentlemen as guests for dinner. This was accordingly held on Monday evening, May 10, at the Fior d'Italia Cafe, in Oakland, and was well attended. The following program will give some idea of the spirit of fun that reigned.

SPECIAL MEETING

Purpose: Special meeting called for Monday, May 10, by members of the steel treaters class to put to a practical test the gastric capacities of our extremely capable and venerable instructors.

Cause: A desire of the above mentioned members to express their appreciation for the able efforts of the above mentioned instructors.

Anticipated Effect: To augment the friendly relations that have existed throughout the course and to create garlic breaths which may require that the remainder of the night be spent in the basement or the garage.

Desired Net Result: An entirely normalized condition not later than Tuesday A. M.

AN OUTLINE OF THE COURSES

Preheater for cast iron stomachs.

Q (queer) alloy garnished with unprecipitated pig with slag and other foreign inclusions

Minced eutectics cased in a white, low carbon, odorless, tasteless, amorphous, pulverulent, homogenous compound insoluble in water, alcohol (including red eye) and other fluids including pancreatic juice obtained from behind the peritoneum.

An air toughened, fatigue resisting substance. By products used for bone charcoal.

Gobs of Enduro drawn back in oil for toughness.

Case hardened butts, decarbonized tips, drawn in butter.

Low melting alloy.

Special quenching solutions to make high speed steel treaters tough.

M. T. Garlinger presided, in view of his size and it being necessary to combine the function of chairman and sergeant at arms at one and the same time. He was ably assisted by S. R. Thurston and L. M. Osborn.

Incidentally, a very thorough questionnaire was given to the instructors whose answers were unanimously credited as minus 90.

Our esteemed member, "Barney" Google, gave a mathematical demonstration that was one of the high points of the evening's amusement. The meeting closed with songs in honor of the instructors. *D. Hanson Grubb.*

The annual business meeting of the Golden Gate Chapter, A. S. S. T., held on May 12, at the Hotel White Cotton, Berkeley. In the absence of the chairman, Frank B. Drake, who is on a visit East, the meeting was presided over by Fred L. Wight and, promptly at the end of dinner, was opened by a roll call.

Mr. Wight then called upon Mr. Moody, chairman of the educational committee, for a report of the course in Ferreous Metallurgy just completed. Mr. Moody was able to give a very satisfactory account of this fifteen weeks' work, both from the standpoint of attendance and of results accomplished, and of the most excellent spirit of thoughtfulness and co-operation shown by the students. S. R. Thurston then described the dinner held on Monday, the 10th, given by the students in honor of the instructors of the course, as a mark of their appreciation, and a final wind-up before disbanding.

The report of the nominating committee followed. J. S. Fowler, chairman, read a letter from our national president, R. M. Bird, with a number of instructive suggestions, many of which applied directly to the choice of names as worked out by the nominating committee. Full weight had naturally to be given to many vital factors. Practical men in daily contact with heat treating problems are as necessary as technical men whose experience is in both shop and laboratory. Executives are needed for organization problems and for the publicity given by the prestige of their positions to aid in the accomplishment of our plans. The necessary element of contact is supplied by salesmen, who are one of the best sources of new members and of publicity. Furthermore, our chapter has received such cordial backing by a number of sustaining members, that recognition must be given to them as fully as possible, and, finally, the candidates have been picked with careful thought of our plans for the future, at present still in a nebulous condition but slowly working toward a definite scheme.

The slate presented follows:

Chairman .....Frank B. Drake, President Johnson Gear Co.
Vice-Chairman .....Welton J. Crook, Associate Professor Stanford University, and Consulting Metallurgist Pacific Coast Steel Co.
Secretary .....C. R. Owens, Industrial Heating Specialist, General Electric Company.

Executive Committee

S. C. Alexander, Asst. Purchasing Agent, Cal.-Hawaiian Sugar Ref. Corp.
Wm. Cohn, Asst. Gen. Supt., Columbia Steel Corp.

Harold E. Gray, Bethlehem Steel Co.
J. H. Gumz, Industrial Engineer, Pacific Gas and Electric Co.
Cecil Hawley, Superintendent, American Forge Co.
E. H. Kottnauer, Metallurgist.
A. Naish, Heat Treat Foreman, Caterpillar Tractor Co.
Fred L. Wight, U. S. Steel Products.
J. H. Wilson, General Manager, Pacific Sheet Steel Co.
Roy T. Wise, General Manager, Standard Die and Tool Co.

It was moved, seconded, and carried that the ticket as presented be unanimously approved. Most of the candidates being present, they were individually introduced. It was a real satisfaction to all the members to know that we had Mr. Drake again at the helm for another year, notwithstanding his multifarious other duties, and an even more successful year is looked forward to on that account alone.

The speaker of the evening needed no introduction. Fred Pelle, of the E. F. Houghton Co., gave an illustrated talk on "CYANIDING AND GAS CARBURIZING" which was followed by an animated and interested discussion.

Most of us have come to look upon an election as rather a bore, from our past experience with other organizations, but it seems as though election nights with Golden Gate Chapter are unusually full of fun. *D. Hanson Grubb.*

HARTFORD CHAPTER

The May meeting of the Hartford Chapter was held Tuesday evening, May 11, 1926, in the Assembly Hall of the Hartford Electric Light Company. The speaker for the evening was G. E. Harcke, director of the Airco Bournonville Welding Institute, who gave a very interesting and instructive talk on oxyacetylene welding and cutting. Mr. Harcke also gave a demonstration of liquid oxygen, showing by immersing various common articles the nearly instantaneous freezing effect of this liquid.

The motion picture film, "Oxygen—The Wonder Worker," was shown illustrating how oxygen is extracted from the air, how acetylene is made, and how both gases are used by large manufacturers in mechanically cutting intricate shapes very rapidly and in automatically welding production parts.

The annual election of officers for the following year was held, which resulted as follows:

Chairman, Dr. Raymond W. Woodward, The Stanley P. Rockwell Company.
Vice-Chairman, D. A. Nemser, Pratt & Whitney Company.
Secretary-Treasurer, Henry I. Moore, The Firth-Sterling Steel Company.
Executive Committee: Edgar D. Lambert, Underwood Typewriter Co.; Henry J. Fishbeck, Pratt & Whitney Co.; Stanley P. Rockwell, The Stanley P. Rockwell Co.; Albert Vuilleumier, New Departure Mfg. Co.; R. F. V. Stanton, Pratt & Whitney Co.

The executive committee will later be enlarged by the election of further members by the above officers.

LEHIGH VALLEY CHAPTER

Dr. Paul D. Merica, director of research, International Nickel Company

of New York City, gave a talk on "Nickel in Cast Iron" at the meeting of members of Lehigh Valley Chapter held on May 7. A short abstract of the paper follows:

There is no question of the fact that iron foundries are today alive to the necessity as well as to the possibilities of improving gray iron. The use of the electric furnace, the development of such a product as the so-called "pearlitic iron" and the use of alloys, particularly nickel and chromium, in this country are evidences of that fact.

Nickel, when added to gray iron, exercises a number of interesting and unusual effects, such as refining the grain, increasing the hardness, decreasing the chill, etc., but if there is any one thing that distinguishes it from the effect of other elements upon gray iron as well as from other methods of improving it, it is that by its use the hardness and strength of gray iron may be increased with a simultaneous decrease of chilling tendency as well as of combined carbon. It is well known that the ordinary methods of increasing strength and hardness of cast iron produce an increase in the chilling tendency of the iron, as well as its combined carbon, leading ultimately to difficulties in machining. This peculiar combination of effects caused by the introduction of nickel means in practice that very much higher values of "machinable hardness" can be obtained by the use of nickel than, as far as is known, by other means.

The use of chromium in conjunction with nickel is recommended for a great many applications. Although chromium functions in iron in quite a different manner than nickel, the two effects supplement each other in many cases very nicely and produce effects which in some cases are superior to those of either alone.

The industrial application of nickel in gray iron is of comparatively recent origin, dating back perhaps to 1918, but is continually growing, the principal applications being those in which greater machinable hardness or strength is required, such as automobile cylinders, cams, gears, etc., or in which additional toughness and freedom from fragility is necessary; for example, in thin section resistance grids.

LOS ANGELES CHAPTER

A special meeting of the Los Angeles Chapter of the American Society for Steel Treating was held in the banquet rooms of the L. A. Creamery, May 7, 1926, at which thirty-eight members and friends were present.

After the dinner various committee reports were read and much interest was taken in a letter received from Prof. Clapp, M. E. of the California Institute of Technology, regarding a Practical Course in Metallurgy and Heat Treating which he would sponsor. Favorable action was taken by the chapter to assist Prof. Clapp in getting this course started for members of the local chapter.

The speaker of the evening, Mr. Pafenbach, metallurgist for the Simons Saw & Knife Company, was then introduced. Mr. Pafenbach chose as his subject, "High Speed Steel," and he, being an authority on the subject, his talk proved to be be one of the most interesting the chapter had heard for some time. A very interesting discussion followed his remarks. Being so

far west, it is a rare treat to have such men as Mr. Pafenbach drop in on us and give our chapter a talk while they are out here on a hurried business trip. We hope more of the large Eastern manufacturers will consider this while their representatives are here in the west.

The regular monthly meeting of the chapter was held in the L. A. Creamery banquet rooms at 6:30 p. m., May 20, 1926. This was a joint meeting of our local chapter and the local chapter of the American Welding Society. Dinner was served to fifty-six members of both societies present. A short business meeting preceded the regular program.

L. G. Hicks of the Standard Oil Company, chairman of the local chapter of the Welding Society, spoke briefly regarding the activities and scope of their organization and then introduced Mr. Stoody of the Stoody Manufacturing Company, inventor of the Stoody Welding Rod. Mr. Stoody spoke on the "Welding of Heat Treated Parts and Heat Treatment of Welded Parts." He also spoke briefly of his experience regarding the welding of cast iron.

E. J. Cipperley of the General Electric Company put on a film entitled the "Wizardry of Wireless," which proved to be very interesting.

Loui Ardouin, secretary of the local chapter of the welding society, together with William Farrar and James Barton, entertained all present with a 2-act comedy, entitled, "Putting Over a Sale," which went over with a bang! The meeting adjourned at 10:30 p. m. *E. C. Black.*

NEW HAVEN CHAPTER

The regular monthly meeting of the New Haven Chapter was held on the evening of the 24th of May at the Hammond Laboratory of Yale University.

A dinner was served at the Hotel Bishop at 6:30 p. m., which was well attended. During the dinner, the usual witty remarks were passed and A. H. d'Arcambal told us of a mew cat at his house by the name of Sultan. Of course, the name is characteristic, so if you have a cat and want a good name for it, ask Darc and he will tell you the rest.

Promptly at 8:00 o'clock, Chairman Brooks called the meeting to order and proceeded with the business of the meeting. W. H. Duley, chairman of the nominating Committee, read the list of officers as selected by the committee, and after a short discussion, the following were unanimously elected and will be installed into office at the June meeting.

Chairman—C. J. Sauer, Bridgeport
Vice-Chairman—W. P. Eddy, Jr., New Haven
Secretary—F. E. Stockwell, New Haven
Treasurer—W. G. Aurand, Wallingford
Executive—J. Brooks, East Haven; L. W. Burns, Bridgeport; W. H. Duley, New Haven; R. G. Hall, Wallingford; H. G. Keshian, Waterbury; R. S. Young, Shelton.

Following the business meeting the chairman introduced the first speaker of the evening, W. P. Kirk, general sales manager of the Pratt and Whitney Company. Mr. Kirk, who is familiar with the business trends of machine tools, gave a very interesting talk accompanied by slides which showed the actual developments of machinery for the past fifty years.

Following this, A. H. d'Arcambal, consulting metallurgist of the Pratt and Whitney Company, who is more familiar with the metallurgical end, described the way alloy steels have been used to solve these problems.

The two speakers very ably covered the field and a very interesting discussion followed which lasted until 11:30 p. m.

The meeting was well attended and we were indeed gratified to see some new faces in our midst.

The annual dinner of the New Haven Chapter will be held at Savin Rock, Friday, June 18. If you have never attended an annual dinner at New Haven, you have missed something, so plan to be on hand this year. *W. G. Aurand.*

NEW YORK CHAPTER

The last meeting of the 1925-26 year for the New York Chapter was held in the rooms of the Merchants' Association, Woolworth Bldg., Wednesday evening, May 19.

The tellers, after canvassing the ballots, reported the following gentlemen as having been elected as officers for the year 1926-27:

For Chairman: Ancel St. John, Consulting Engineer.

For Vice-Chairman: Charles McKnight, Jr., International Nickel Co.

For Secretary-Treasurer: T. N. Holden, Jr., E. W. Bliss Co.

Executive Committee:

Edgar C. Bain, Carbide and Carbon Laboratories
E. P. Gaffney, Crucible Steel Co. of America
J. J. Crowe, Air Reduction Sales Co.
R. L. Johnston, Aluminum Die Castings Co.
Francis F. Lucas, Western Electric Co.
Charles von Schlutter, Firth-Sterling Steel Co.
John O. Rinek, Crucible Steel Co. of America

As a fitting climax of a well-arranged series of meetings on the manufacture and uses of fine steels, J. L. McCloud from the Central Metallurgical Laboratory of the Ford Motor Company at Dearborn, Mich., discussed "Steel Selection for Automotive Work." Unfortunately, the sudden warm spell reduced the attendance somewhat below the average, but every person present was keenly interested in the paper, and many joined in an extended discussion following it.

Mr. McCloud's paper was so replete with intimate detail, that it is almost impossible to abstract. Furthermore, it is of such value that it will undoubtedly be used before other chapters and eventually will find its way entire into the "TRANSACTIONS". Mr. McCloud confessed that although he has been very active in the Steel Treaters' Society, he has never since the first general meeting at Philadelphia, presented a formal paper before any group. In view of his excellent address, the New York Chapter feels proud of its oft-demonstrated ability to dig up unusually good speakers on little discussed phases of the subject. Not that the selection of steel for automotive work has never before been described before steel treaters, but seldom before with this background of a wealth of experience and applications to light automobiles, heavy expensive cars, tractors, railroad rolling stock and aircraft.

It is especially worthy of note that the tendency of the Ford organiza-

tion is to heat-treat straight carbon steels and to rely very largely upon straight chromium steels for most of the applications to which a wide variety of analyses have been recommended and used elsewhere. This practice is possible to the Ford Motor Company with their ability to purchase in large quantity, to segregate the shipments closely by heat analyses, and their accurate and automatic control of the various hardening-room operations.

Mr. McCloud described briefly two or three interesting applications of heat-treated steel to locomotives operating on the Detroit, Toledo & Ironton Railroad. They are especially gratified by the performance of heat-treated alloy steel forgings for connecting rods and side rods on the main drivers.

In this connection, it may be remembered that one of the large English railway systems equipped a number of railroads with heat-treated alloy steel rods, and issued the most rigid instructions that repairs on such rods were to be made only in one shop where they had the necessary personnel and furnaces for reheat-treatment. However, this rule was impossible to enforce and some serious accidents were narrowly averted from broken rods, each of which it was discovered had been heated for straightening or for some other purpose at an outlying repair shop. For this reason it was reported that all alloy steel side rods were removed from the engines.

Perhaps the discipline of Mr. Ford's railroad is better than on the larger English system referred to, and perhaps Mr. Ford will be able to confine all engine repairs to shops properly equipped to handle alloy steels. At any rate, this particular development will unquestionably be watched with interest by American railway engineers. *E. E. Thum.*

PHILADELPHIA CHAPTER

The cooperative evening course in Heat Treatment and Metallography of Steel, conducted at Temple University under the auspices of the Philadelphia Chapter American Society for Steel Treating, closed its fifth session with an exhibit of the laboratory and students' work, on the evenings of May 26 and 27. There was a capacity attendance of visitors both evenings.

The complete laboratory equipment was demonstrated by the students, including Bausch & Lomb and Leitz metallurgical microscopes; Leeds & Northrup, Brown, Thwing, Wilson-Maeulen and Republic pyrometers; Olsen, Brinell, Shore scleroscope and Rockwell hardness testers; Leeds & Northrup and Hevi-Duty electric furnaces; American Gas Furnace Co. furnaces for carbon and high speed steel; carburizing machine; pots for lead or salt baths; bench forges; dark room; Warner & Swasey polishing machine, etc.

Reports by the students covering investigations of practical problems in steel treatment were shown.

The enrollment has grown from 13 in the first year, to 70 in the present year, a total of 203 students having attended during the five years. One student, Howard Paget, came from California to attend the class. Nearly all of them are employed with leading industries in the Philadelphia district.

The students are of all ages and occupations, including heat-treaters, machinists, inspectors, clerks, foremen, salesmen, purchasing agents, engineers, teachers, managers and executives.

This season marked the introduction of the advanced class, consisting of

men who have completed the first year's work, or who have had equivalent training or experience elsewhere.

The course consists of a series of 30 lectures and 30 laboratory periods, beginning in October and ending in May. The lectures cover the fundamental principles of the subject from the manufacture of steel to the treatment and testing of the finished tool or article of manufacture. In the laboratory the students learn the use of the equipment used in steel treating and testing, and apply the principles covered in the lectures. A special text has been written for the course.

In addition to the director, there are four laboratory instructors, and ten special lecturers, including Prof. Bradley Stoughton, Dr. Geo. L. Kelley, Dr. W. M. Mitchell, J. J. Crowe, Arthur Green, D. K. Bullens, A. L. Collins, J. E. Chapin, Theo. Wiedemann, and Dr. H. F Moore. All of the teaching staff are members of the Society, and are actively engaged in some phase of steel treating.

The course is self-supporting, but is not run for profit. Its success is a notable example of the results of wholehearted cooperation in education and industry.

H. C. Knerr.

PITTSBURGH CHAPTER

At a specially called meeting, held on the evening of April 15, the Pittsburgh Chapter met at the Carnegie Institute of Technology to hear F. F. Lucas of the Bell Telephone Research Laboratories.

Mr. Lucas's talk was indeed a treat, especially to the metallographers.

In his opening remarks he told of the work of E. Hooke of England, which was done about three hundred years ago and which could not help but induce admiration when the crude instruments and equipment used, were described.

Mr. Lucas then gave a detailed description of his laboratory and its equipment. In telling of the preparation of the specimens for the microscope he advised the use of magnesium oxide free of carbonates on a wheel covered with broadcloth for the final polishing. The grade or brand of broadcloth used is called "Kitten's Ear."

This was followed by a description with pictures of the crystallization of iron and photomicrographs at high magnification with different etchings. Austenite, martensite, troostite, sorbite and pearlite were illustrated and accompanied by remarks as to their probable formation. An interesting notation at this point was that martensite needles never cross a grain boundary nor a twinning plane.

Mr. Lucas told of his experience with the ultra violet ray equipment which he has been pioneering. The difficulties of operation and how they were overcome were enumerated and photomicrographs made by this method were shown.

At the conclusion of this talk the speaker was asked many questions pertaining to metallography which aroused considerable interest.

Harry A. Neeb, Jr.

RHODE ISLAND CHAPTER

On the evening of February 24, 1926, the Rhode Island Chapter, American

Society for Steel Treating and the local section of American Society of Mechanical Engineers held a joint meeting at the Providence Engineering Society Rooms at which S. Fitzsimmons presided.

Peculiarly fitted to the appetite of a mixed group such as was present at this meeting, the caterers on the program committee had as the "piece de resistance" of the menu, "The Story of a Bar of Tool Steel."

Through the medium of a splendid series of motion pictures, the assembly was transported to Bethlehem, Pennsylvania, and under the able guidance of G. A. Richardson, publicity manager for the Bethlehem Steel Company, were allowed the privilege of making a tour of inspection of the various departments of the vast plant, devoted to the making of tool steel.

Beginning with the new materials the audience was shown how the various ingredients, composing the heat, were carefully weighed and at the proper time added to the electric furnace. At various intervals during the heat samples were taken for inspection, and indeed, throughout the whole process the number and kinds of inspections and tests made were a revelation to a great many present.

During the melting down period, Mr. Richardson gave a very concise description of the chemistry of the process.

When the melting down had been completed, we were shown how the molten steel was teemed into the ingot molds and from the ingot, through the various mechanical processes of forging, rolling, etc., thence to the annealing, through the many and various inspections and finally to the shipping platform of the mill.

All through the numerous reels, the pictures were made doubly interesting by Mr. Richardson's explanation, bringing out as he did, the most interesting features of the pictures.

As an extra treat, a visit was made to the laboratories of the company and there we found chemical and physical testing applied to the nth power, for we saw routine chemical and physical tests familiar to most of us, other unfamiliar tests read about by some of us and still other tests not heard about by any of us.

But we do believe that as a result of the pictures and Mr. Richardson's discourse, the majority of those present will have a little more respect for that unromantic looking bar of steel that comes into the plant, giving no clue, by its appearance, of the great amount of care expended in its behalf, all the way from the raw materials to the mill shipping platform. *E. J. Sullivan.*

ST. LOUIS CHAPTER

The 58th monthly meeting of the St. Louis Chapter was held on Friday evening, May 21, at the American Annex Hotel. A good crowd was on hand to hear C. P. Richter, metallurgical engineer of the Central Steel Co., Massillon, Ohio. The subject of Mr. Richter's address was "Factors Governing the Selection of an Alloy Steel." It was well illustrated with a number of most interesting lantern slides.

Mr. Richter's talk appealed to the interested listeners as something they all wanted to know; in clear, plain language the address seemed to reach every one and clear up some of the doubts concerning the nature of the many

alloys and combinations of alloys met with in general use today. Mr. Richter showed how some alloys come to the front and, as knowledge widens, are superseded by still others.

First of all the chief combinations of alloys were enumerated, each combination having qualities for certain purposes. It is well known that nickel reduces scale, that chromium adds hardness, that vanadium is a cleanser and adds toughness, that tungsten gives the quality of red hardness. Of all the various combination of alloys, chromium-vanadium is the most generally used. Mr. Richter divided the alloys up into various types, such as carburizing types, forging types, gear steels, etc. For the carburizing types, carbon and nickel steels are used; forging steels are both water hardening and oil hardening, chromium-vanadium steel quenched in a salt bath is satisfactory for gears.

Of the newer combinations, nickel-molybdenum is coming into its own. The main point that struck the members was the bright future of alloy steel, the endless possibilities, the great progress in the last ten years. Alloys are tools in the engineer's and metallurgist's hands that can be used to tackle each engineering difficulty as it comes along.

At the annual election of officers the following were elected to serve in 1926-1927:

Chairman, C. B. Swander, Wagner Electric Corporation.

Vice-Chairman, J. J. Bowden, Laclede Steel Company.

Secretary-Treasurer, C. G. Werscheid, Crucible Steel Company.

A hearty vote of thanks was given Mr. Richter for his excellent address, and the meeting adjourned. *Alan Jackman.*

WASHINGTON CHAPTER

At the last monthly meeting of the 1925-1926 season, on May 21, the membership instructed the secretary to cast a unanimous ballot in favor of the officers selected by the nominating committee, to serve for the ensuing year. The officers so chosen are as follows:

Chairman, O. Z. Klopsch.

Vice-Chairman, Emil Gathmann.

Secretary-Treasurer, G. W. Quick.

Chairman Meeting Committee, Samuel Epstein.

Members Executive Committee: P. E. McKinney, J. E. Crown, Jerome Strauss, J. W. Barnett, and G. M. Nauss.

Following this election, Dr. H. W. Gillett presented an illustrated talk on the subject of "The Behaviour of Steel Under Repeated Stress." Enough knowledge is available concerning the static properties of metals so that modern structures do not fail for lack of strength as a result of the application of slowly applied loads. Those unexpected failures that do occur can be ascribed to either corrosion or wear or fatigue or a combination of these causes; of these failures the greater number are the result of the effects of alternations of stress. Since only an extremely small percentage of finished parts of a given type fail as a result of fatigue, it appears that designs are generally suited to the purposes to which they are applied, and inquiries have therefore been directed to why the small percentage should fail. Dr. Gillett

took occasion to strike again at that old idea of many members of the engineering profession that breakages of this type are due to "crystallization." That metals are naturally crystalline has not been fully understood by the profession and they refrain from attaching any significance to the occurrence, in a member containing a fracture by fatigue with no deformation, of silky surfaces on subsequent rupture accompanied by much deformation.

Most fatigue failures originate at the surface of steel parts due to the greater number being subjected to repeated bending stress; with such loading, the maximum unit stress is at the surface. Some failures of the progressive type have their origin within the metal mass, a notable example being transverse fissures in rails; it is not, however, generally agreed that these fractures are the result of fatigue. In any case, the origin of the fatigue failure is at a notch of some type crossing the line of applied stress and thereby introducing severe stress concentration; this origin may be a nick, a tool mark, a sharp fillet, a non-metallic inclusion, etc.

In reviewing the history of the study of fatigue it appears that early workers did not appreciate the existence of an endurance limit or stress below which no number of load applications will cause failure. Recent work has definitely proved the existence of such a stress in steels but not positively in the non-ferrous metals. It is important to note that very little work has been carried beyond a billion cycles of stress and that many parts, particularly turbine shafts, have an actual life exceeding this.

In referring to experimental details in the determination of the behaviour of metals when subject to alternating stress, Dr. Gillett remarked that the speed of testing up to 2,000 cycles per minute had been found not to affect the results. In very soft steels the endurance limit would usually be found at 20 million cycles or less and in the harder structural steels frequently as low as ½ million cycles. The results obtained by plotting the data on simple Cartesian coordinates and on a logarithmic or semi-logarithmic scale were discussed and the advantages of the logarithmic method explained. Theoretically some question arises as to whether the logarithmic curve (or S-N diagram) beyond the endurance limit is actually horizontal or dips very slightly. Any inclination, if present, cannot be great, and there is, therefore, at least an "engineering endurance limit."

In all of the work thus far recorded there will be found some results which are below the curve for any given material, and thus one may say that there are two endurance limits, one for each material taken as a whole, and another for each individual specimen. A wide scatter of points has frequently occurred in materials which have been known to be "dirty."

The number of cycles at which the endurance limit is generally found has been previously noted. If endurance tests upon a series of iron alloys of various strengths are plotted on the same coordinates, the curve connecting the "knees" of the various graphs is practically a straight line thus indicating a relationship between the strength or endurance limit of the steel and the number of cycles at which the endurance limit is found. It is also noteworthy that the harder the steel the more marked is the bend in the S-N curve.

Earlier investigations had all shown a difference in the endurance limit according to whether it had been determined by axial tension and compression or by alternation of stress in bending, the former value being always lower. In a recent series of very carefully conducted experiments, Irwin has shown that the values determined by the two methods are the same for any steel, as indeed it would seem that it should be. It has, therefore, been proven that the earlier results were in error probably due to the great difficulty in securing truly axial loading when dealing with stress alternating from tension to compression. Many engineering structures are no doubt subject to uncertainties similar to those of the early tests as actual stresses may be quite different in both magnitude and direction from those determined by the usual engineering analysis.

Dr. Gillett referred to determinations of the endurance limit of materials under conditions in which the mean stress of the range was not zero. Most of the results obtained have not possessed a great degree of reliability but recent work of McAdam in torsion shows that the life of the specimen under a given stress range is the same regardless of the mean stress applied when that mean stress and the range of stress are such that the elastic limit, and in some cases the yield point, is not exceeded. It does not matter if the endurance limit for stresses equally distributed about zero is exceeded by the stress used in any unbalanced cycle provided the condition previously noted is adhered to.

The ratio between endurance limit and tensile strength in steels up to about 175,000 pounds per square inch tensile strength averages 0.5, i. e. the endurance range equals the tensile strength. Some values are obtained slightly above, some slightly below this average, but it affords at least a good first-approximation. This figure applies to both carbon and alloy steels so that insofar as endurance limit is concerned, "steel is steel." In general, fine grained materials tend to give values slightly higher than the mean and coarse grained materials slightly lower, but this is not true in every instance. The ratio of endurance limit to tensile strength appears to be quite accidental, for no relationship has thus far been found between endurance limit and any true physical property of the material. Above 175,000 pounds per square inch tensile strength the trend is toward a decrease in the ratio. Most of the low values that have been found, however, are in the cerium-bearing steels, and these are not only "dirty" but variably "dirty." Unfortunately not much work has been done upon these harder structural steels so that statements with regard to this ratio cannot be positive. Many attempts have been made to try to prove that in these "dirty" steels the origin of failure is an internal notch caused by an inclusion or group of inclusions, but definite proof is lacking. In plates, transverse tests give lower values than do longitudinal tests with the difference being greater for the "dirty" steels. The lowering of the ratio on the harder steels may be partly due to the greater effect of sonims in these steels, but it is also due in greater measure to another feature which is described later.

It was shown that an increase in the endurance limit may result from the application of alternating stress slightly below the normal endurance limit

followed by an increase in the stress range; i. e. strengthening occurs by under-stressing. Under-stressing of heat treated steels combined with re-tempering at the same or a higher tempering temperature (with frequently but not always a decrease in hardness resulting) also produced strengthening in the nature of an increase in endurance limit. These results point toward the existence of much internal stress in these steels when tempering for a length of time generally regarded as ample. They indicate further that this normal tempering time is too short and probably is the greatest factor in lowering the ratio of endurance limit to tensile strength in the harder steels. Further work along these lines may show that the ratio is constant even up to very high tensile strengths.

The endurance limit may in some cases be much above the static proportional limit, and in these cases the latter value may not be a true one, i. e. its location may be the result of internal stresses. In soft metals, it is not so simple to explain why this phenomenon should occur.

Thus, the endurance limit is the result of a balance reached between two factors, one the deteriorating effect of applied stress and the other the strengthening effect of this same stress. Steel is not isotropic but is composed of a large number of small differently oriented crystals. Stress as usually computed is merely an integrated value over all of these minute units. It is, therefore, possible that the endurance limit is the elastic limit or the yield point of the most disadvantageously located grain. That grain which first slips is strengthened by cold working and a stress readjustment occurs which causes the next slipping to take place on some neighboring grain since the previously strengthened one is more resistant to additional slipping. Consequently each grain in turn grows stronger, but once a crack has been started by a load above that corresponding to the endurance limit of the bar, great stress concentration occurs at this infinitely sharp submicroscopic crack and the material is permanently injured; the crack then spreads rapidly. Most questions arising in endurance work can be answered by a consideration of microscopic or submicroscopic features of the metal rather than the gross features.

Local concentration of stress, either as a result of surface condition or inclusions or specimen design will lower the endurance limit, and the harder the steel the more sensitive is it to these conditions. However, the effect of the notch in actual practice is not so great in magnitude as is generally indicated by mathematical analysis. Failure of the very small percentage of properly designed parts is usually due to these individual stress-raising local conditions; this is the reply to the query noted earlier in this report.

Dr. Gillett referred briefly to the rapid methods of endurance testing—in particular the measurement of temperature rise to determine the stress at which sudden change in this rise occurs and also the method of determining that stress which produces a sudden change in the deflection of a rotating cantilever specimen. On hard steels, and many nonferrous metals that show a particularly low endurance limit by the usual methods, these rapid methods give extremely high values. According to Gough, in these rapid methods time is not available for surface or internal notches to be effective. Conse-

quently, any effort to employ these rapid methods in inspection for the purpose of locating material internally defective would not be reliable.

In the discussion that followed this paper, Dr. L. B. Tuckerman expressed the opinion of the engineer on the results of endurance testing. The engineer in order to function must produce concrete accomplishment, and in doing this uses the best available knowledge. It is true that little is available on the correlation of the strength properties of test specimens with those of the masses of metal actually employed in construction. This offers a serious handicap in design which is too often totally disregarded; it applies to both static and dynamic strength data. It is indeed unfortunate that the usual basis of design are so purely conventional and that many designs are the result of comparison with studies of previous similar structures. But on truly new designs even this information is lacking and this feature oftentimes accounts for failures. Dr. Tuckerman considered the scatter of points in fatigue test results as being due to one of two causes,—the degree of reliability of the material and the design of the test specimen combined with a lack of knowledge of the true stress due to the testing method employed. Of these the second cause exerted the greater influence. The effect of heterogeneity due to interplay of atomic forces even in the absence of crystalline structure, varied orientation and included particles was of great importance. Because of it, stress can never be truly uniform and this must be borne in mind in developing improvements in testing and also in the analysis of the nature of fatigue. From the engineer's viewpoint, for static loading design should insure freedom from the possibility of movement beyond the elastic range. In dynamic loading two types must be considered—repeated permanent deformation and repetition of stresses that did not produce permanent deformation. The latter occurred not only in crank-shafts, turbine shafts and the like, but anywhere that vibratory movement was found, even though the vibration might be accidental.

Referring to the non-ferrous metals, Dr. McAdam mentioned that some of the yielded S-N curves which seemed to have a horizontal asymptote representing the endurance limit. Actually running a test at a billion or more cycles was time-consuming, but it was possible to secure sufficient data by extrapolation if the portion of the curve in the higher stress ranges were more definitely known. In these ranges, however, testing developed considerable heat, but by the use of water-cooled specimens the earlier portion of the curve could be obtained and extrapolation then rendered more certain. Dr. McAdam considered the scatter of points to be due more to non-uniformity of the material than to the method of testing (referring particularly to the rotating beam method).

Many other interesting features were brought up in the course of the discussion, but space prohibits recording of further details. *Jerome Strauss.*

FORT WAYNE GROUP

The third regular meeting of the Fort Wayne Group was held on June 9, with F. R. Palmer of the Carpenter Steel Company as speaker. He told the boys some things about steel tools and tool steels that they had never

(Continued on Page 172)

NEW MEMBERS OF THE SOCIETY

EXPLANATION OF ABBREVIATIONS. M represents Member; A represents Associate Member; S represents Sustaining Member; J represents Junior Member; and Sub represents Subscribing Member. The figure following the letter shows the month in which the membership became effective.

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News of the Chapters

(Continued from Page 163)

heard before. The crowd may not have been as big as some they have in Chicago or Detroit but the enthusiasm can't be beaten.

The speaker announced that his talk was concerning "steel tools" and not "tool steels." Then followed an excellent lecture on steel tools from the standpoint of the designer, the toolmaker, the hardener, the grinder, and the user. The talk was practical and the membership went away with the belief that the speaker "spoke from the fullness of his wisdom." The members here are getting into the habit of bringing in their steel problems, and when opportunity for questions was given there were plenty of them.

Fred C. Smith.

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SALESMAN. Graduate of mechanical or electrical engineering course, preferably from class of '23 to '26, for service and sales work on Hump and Homo Heat Treating Furnaces. Short training period necessary before active field work. Apply by letter only, giving age, marital state, education, schools attended, experience, etc. State salary expected and enclose photograph. Address—Personnel Dept., Leeds and Northrup Company, 4901 Stenton Ave., Phila., Pa.

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FOR SALE: One Bausch and Lomb Metallurgical Microscope completely equipped for taking photomicrographs. In good condition. Will sell at a low price. Address 6-30.

FOR SALE—One Stewart No. 10 oven furnace for heat treating, annealing and carbonizing; used but a short time; in excellent condition.

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Items of Interest

L. Gerald Firth was appointed general manager of the Firth-Sterling Steel Company and R. S. Stevick, works manager, at a meeting of the board of directors of that company held recently. Mr. Firth was formerly works manager and Mr. Stevick his assistant.

The Philadelphia Rust Proof Company which specializes in the coating of metals to protect them from rust and corrosion by the processes known as Parkerizing, the Udylite Process of cadmium coating and Sherardizing have recently moved and enlarged their plant and are now located at 109 West Montgomery Ave. Any visitors to the sesqui-Centennial who are interested in having rust proofing done are invited to call and inspect their plant.

M. W. Caruthers has taken charge of the Chicago Branch of the Braeburn Alloy Steel Corporation of Braeburn, Pennsylvania. Mr. Caruthers was formerly in the service engineering department of the Crucible Steel Company of America with headquarters at New York.

Allan K. Hamer, member A. S. S. T., formerly metallurgist of Keystone Forge Company, has accepted a similar position with the U. S. Gauge Company of Sellersville, Pa.

The International Nickel Company, 67 Wall Street, New York City, has available Bulletin No. 201 entitled "Properties and Applications of Nickel and Nickel-Chromium Cast Iron." Also Bulletin No. 202 entitled "Improving Gray Cast Iron With Nickel." Sent upon request without obligation.

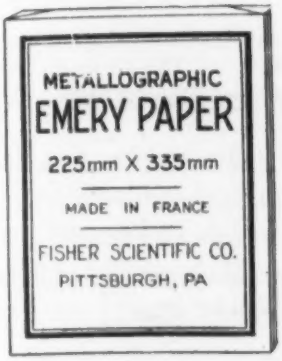
The National Machinery Company of Tiffin, Ohio, has just issued "National Forging Machine Talk No. 57." The subject of this talk is "Forging the Ends of Ladder Rungs." An illustration of the die used is given and the process is described. Gratis upon request.

George R. Hanks has recently been elected vice-president of the Taylor-Wharton Iron and Steel Company, of High Bridge, N. J., to succeed the late Victor Angerer, who died on May 5. He was formerly superintendent and later acted in the capacity of assistant to the president.

Alan Jackman is now manager of the Cleveland sales office of E. S. Jackman and Company, Chicago dealers in tool steels. Mr. Jackman formerly resided in St. Louis and was connected with the St. Louis and Chicago branches of his company.

(Continued on Page 36 Adv. Sec.)

FRENCH METALLOGRAPHIC EMERY PAPER



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Prof. Bradley Stoughton, head of the department of metallurgy of Lehigh University, Bethlehem, Pa., sailed for Europe June 5 on the "Chicago." During his three months' trip abroad Prof. Stoughton will visit the metallurgical departments of several German, British, French and other universities.

F. F. Harter has recently been made branch manager at Philadelphia for the Crucible Steel Company of America, New York. His headquarters are 1224 Callowhill Street, Philadelphia.

T. F. Rose, formerly assistant branch manager of the Chicago office for Timken Roller Bearing Service and Sales Company, Canton, Ohio, has been appointed manager of the Cincinnati branch.

H. C. Sauer has been appointed manager of the Detroit branch. He was formerly assistant branch manager at Cleveland.

Leonidas Merritt, discoverer of the first iron ore at the western end of the Mesaba range, died recently in Duluth at the age of 82.

W. B. Dilley has joined the staff of Production Materials, Ltd., Canadian representatives of E. F. Houghton and Company, Philadelphia, in the capacity of distributor for Toronto.

Dr. Albert Sauveur, professor of metallurgy, Harvard University, Cambridge, expects to sail for Europe on July 3, returning September 8. He plans to attend the meeting of the Institute of Metals (British) at Liege, Belgium, and also the meeting of the Iron and Steel Institute which is to be held in Stockholm, Sweden.

The Los Angeles chapter has instituted a course of metallurgy under the direction of Prof. W. H. Clapp, instructor of metallurgy at the California Institute of Technology.

The response for enrollment is quite heavy and the first meeting was held in the conference room of the Westinghouse Electric and Manufacturing Company at 7:30 p. m. on Tuesday, June 29. The first lecture presented was "Iron and Industrial Progress." Other lectures of the course will follow regularly on each Tuesday evening.

The United States Civil Service Commission announces the following open competitive examination for the position of Metallurgical Laboratorian.

Applications for this position must be on file at Washington, D. C., not later than August 7. The date for assembling of competitors will be stated on the admission cards sent applicants after the close of receipt of applications.

The examination is to fill vacancies at the Naval Academy, Annapolis, Md., at $4.32 a day, and vacancies occurring in the Federal classified field service throughout the United States.

Full information and application blanks may be obtained from the United States Civil Service Commission, Washington, D. C., or the secretary of the board of U. S. civil service examiners at the post office or customhouse in any city.

July

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